

The Journal of the

INSTITUTE OF METALS

and

METALLURGICAL ABSTRACTS



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JANUARY 1948

The Swiss know a lot about



Any third form fag know that not much of Switzerland is on the level. The Swiss know it too. Perhaps that is why the Swiss are so keen on overcoming the power of g (short for gravity) that they have developed their light alloy industry to such an advanced stage. We in this country could save ourselves much wasted energy and expense if we, too, overcame g by a greater use of the light alloys that H.D.A. produce. We have the alloys, we know how to use them — all that is needed is more manufacturers to . . .

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To. Mr. Dir......

Memo. Ref. 124/P.S/47.

Continued.

7.7.47

decided, unanimously, that the best, if not the only way to keep pace with the new production schedule is to use pressure die castings.

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shall we
specify?
M.D.

9.7.47

B.E.H.
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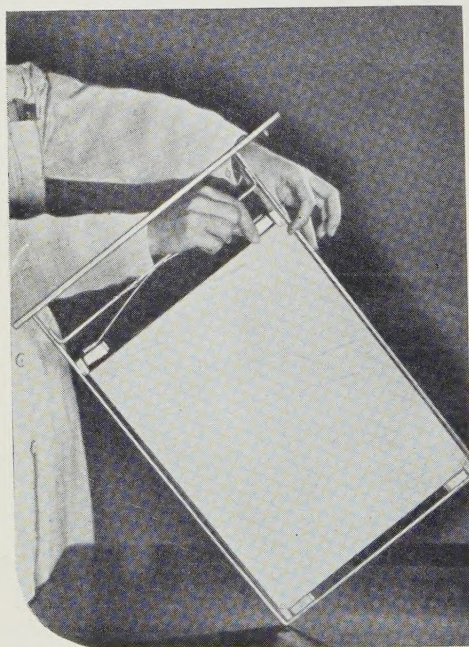
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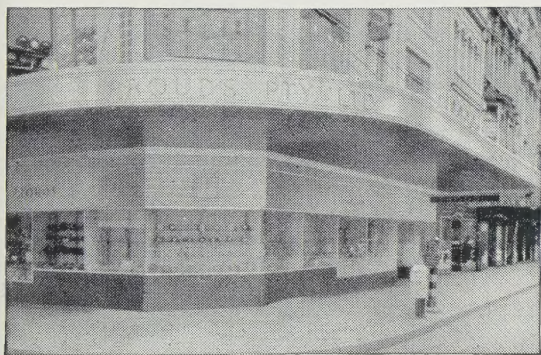
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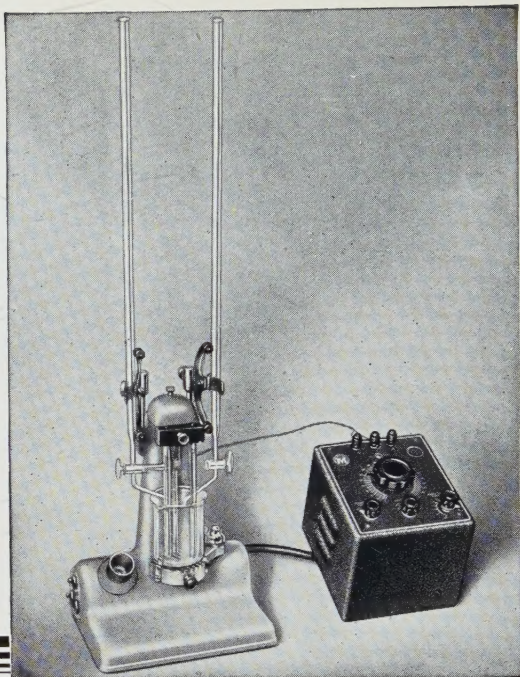
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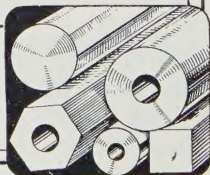
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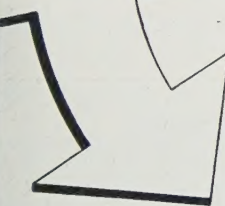
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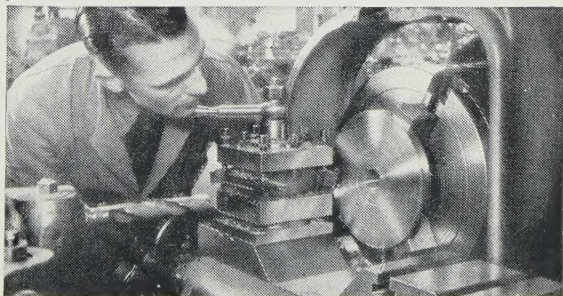


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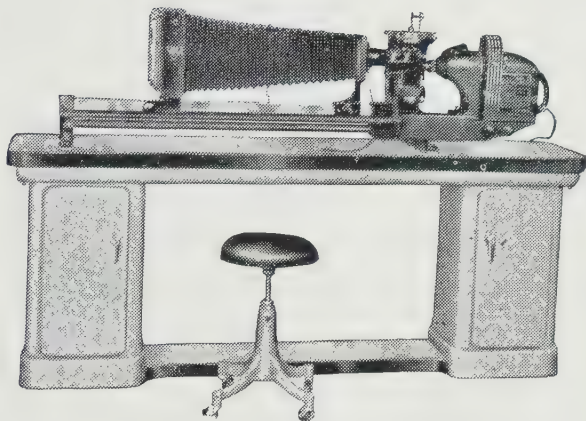
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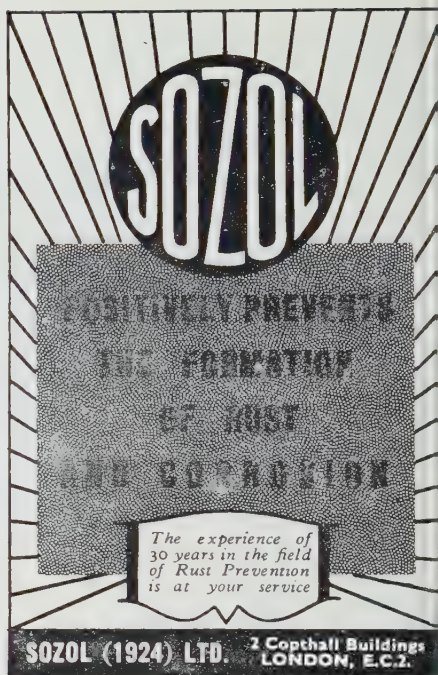
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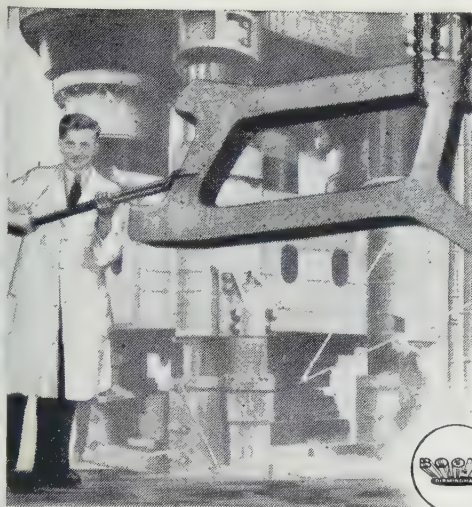
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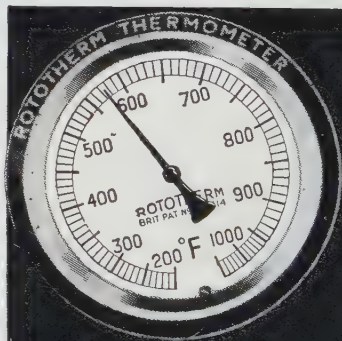
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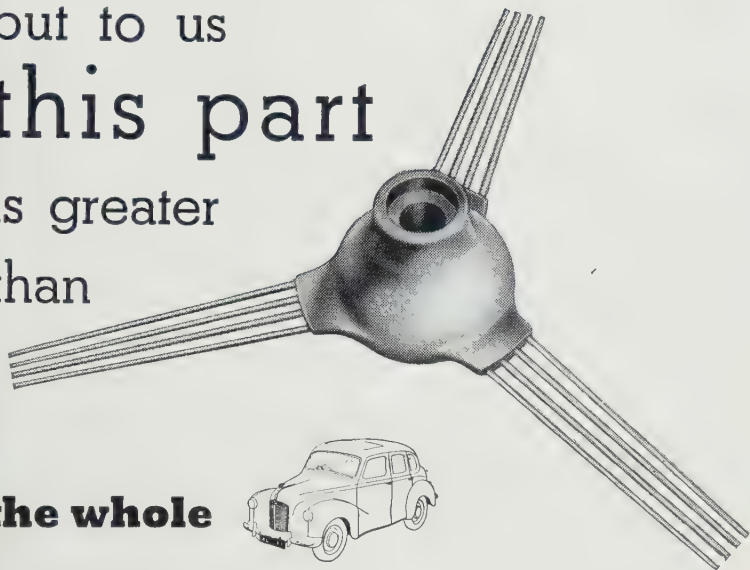
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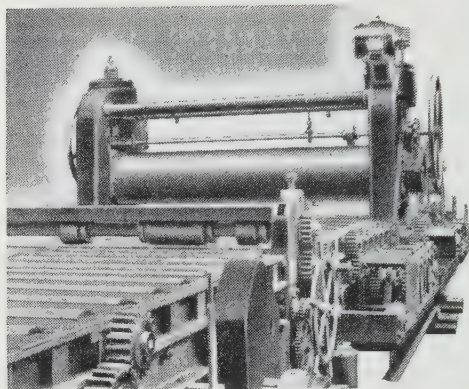
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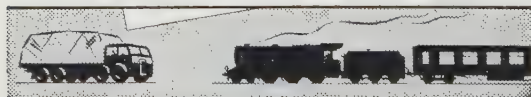
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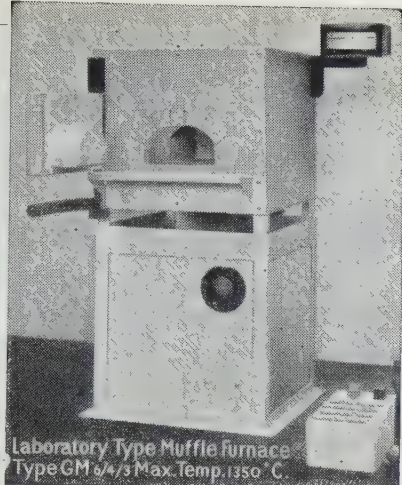
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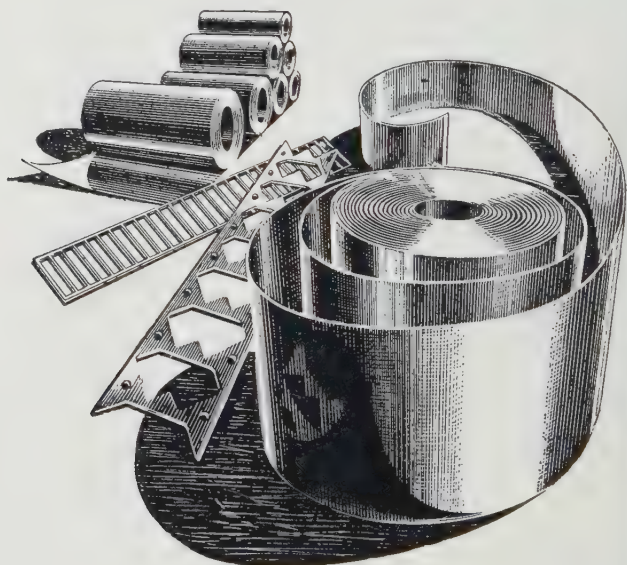
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January 1948

THE INSTITUTE OF METALS

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A NEW YEAR MESSAGE FROM THE PRÉSIDENT

ONCE again at the beginning of a New Year we must possess our souls in patience and renew our trustfulness and courage so that we may overcome the difficulties with which we are faced.

The past twelve months have brought many setbacks to the achievement of recovery at home and stability abroad, but, so far as the Institute is concerned, we can look back on 1947 with some satisfaction.

We have held a number of most successful meetings, notably the Autumn Meeting in Glasgow, we have set our administrative house in order, and we have a programme for 1948 which should enable members to partake in interesting and important discussions and receive regularly a *Journal* which, while maintaining its high standard unimpaired, will have a more direct appeal to our many industrial members than it has had in the past. I hope it will be a medium whereby members may learn what others engaged in the science and practice of Metallurgy are thinking and doing. Thus we shall promote that spirit of constructive criticism and co-operation which it is the main object of the Institute to foster.

My wish to all members is that the New Year will not only bring you happiness in your homes, but joy in your work, and that your membership of the Institute will contribute substantially to the fulfilment of this wish.

P. G. J. Gueterbock

INSTITUTE NEWS AND ANNOUNCEMENTS

INSTITUTE OF METALS (PLATINUM) MEDAL, 1948

The Institute of Metals (Platinum) Medal for 1948 has been awarded to DR. ROBERT CROOKS STANLEY, M.E., in recognition of his outstanding services to the non-ferrous metal industries.

Dr. Stanley will be unable to be in England in March to receive the medal in person at the Annual General Meeting.

ANNUAL GENERAL MEETING, LONDON, 16, 17, AND 18 MARCH 1948

The Annual General Meeting of the Institute will be held on Tuesday, Wednesday, and Thursday, 16, 17, and 18 March 1948, when Sir Arthur Smout, J.P., will take office as the new President. A detailed programme will be sent to all members in due course.

The meeting will commence on the afternoon of Tuesday, 16 March, when the Reports of Council and of the Honorary Treasurer will be presented, and the new President will deliver his Address.

The morning and afternoon of Wednesday, 17 March, will be devoted to the presentation and discussion of papers.

On the morning of Thursday, 18 March, the presentation and discussion of papers will be resumed, and in the afternoon there will be visits to laboratories and works of metallurgical interest, with a special visit arranged for ladies.

The business meetings will be held at the Institution of Civil Engineers, Great George Street, Westminster, S.W.1, while buffet lunches will be served at the Royal Institution of Chartered Surveyors, Great George Street, Westminster, S.W.1, on the Wednesday and Thursday of the meeting, to ticket holders.

It is hoped that there will be a large attendance at this meeting.

MAY LECTURE AND MAY MEETING, 1948

The dates *provisionally* arranged are Thursday, 13 May (May Lecture in the evening), and Friday, 14 May (discussion of papers in the morning).

AUTUMN MEETING, 1948

The Annual Autumn Meeting of the Institute will be held at Cambridge, from Tuesday, 14 September, to Friday, 17 September 1948, inclusive.

MEMBERSHIP

Members who are elected between 1 January and 30 June have the privilege of paying a half-year's subscription to cover that period.

Copies of the Institute's Articles of Association and of the Application Form for Membership may be obtained from the Secretary.

NEWS AND ANNOUNCEMENTS

SECOND-HAND VOLUMES OF THE JOURNAL AND METALLURGICAL ABSTRACTS

A few sets and odd second-hand bound volumes of the *Journal* and *Metallurgical Abstracts* are available for purchase by members and libraries at specially reduced rates. Particulars may be obtained from the Secretary.

LOOSE COPIES OF PAPERS

Copies of papers published by the Institute, from the first published to those of recent years, are available for purchase at the price of 2s. 6d. each, post free (1s. 3d. each, post free, to members).

PERSONALITIES

Professor P. A. J. Chevenard

Pierre Antoine Jean Chevenard, who has been elected an Honorary Member of the Institute of Metals, was born in 1888 at Thizy, a little town in the Département du Rhône, about 30 miles north-west of Lyons. He studied at the Ecole des Mines de Saint-Etienne, from which he graduated top in 1910 and then took up a post as technical secretary at the Decazeville works of the well-known firm of Commentry, Fourchambault et Decazeville, but in January of the following year he was transferred to the laboratory of their steel works at Imphy. Chevenard was the only metallurgist in this embryonic laboratory, which contained, in addition to apparatus for chemical analysis and mechanical testing, one metallurgical microscope and one electrical resistance furnace fitted with a Le Chatelier thermocouple.

M. Chevenard very soon supplemented this meagre equipment with a dilatometer which he designed and made himself. He then carried out work on nickel-chromium steels of such a remarkable character that in the spring of 1912—a little over a year after he had entered the laboratory—M. Fayol, the Director-General of the firm, created a new post for him, that of "engineer-in-charge of metallographic research at Imphy", and his job was, in addition to solving the problems which arose in connection with difficulties of manufacture and the needs of customers, "methodically to study special steels so as to discover and exploit their unusual properties".

In the course of this study, the results of which are well known to all interested in special steels, Chevenard designed and made the very large range of instruments which will always be associated with his name. The first of these was the dilatometer referred to above, in which the principle of the "optical tripod" was employed. Others, many of which also used the optical tripod, included thermomagnetometers, thermostats, machines for testing the mechanical properties of specimens which had a volume of only 1/1000 of the standard test-bars, and many others.

With these instruments Chevenard investigated the effects of the transformations in a very large number of special steels, more especially nickel-steels with or without additions of chromium, tungsten, cobalt, copper, and other elements, over a range of temperature between -195° and 900° C. He further studied the effect of heat-treating these steels on their resistance to stress-

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corrosion, both in liquids and gases, and also on their fatigue properties.

This work led in the first place to an improvement in the products of the Imphy works and later to the development of new materials. One of the earliest of these was the heat-resisting alloy containing 60% of nickel, 10% of chromium, and 4% of tungsten. This was used in making receivers for the manufacture of ammonia under a pressure of 1000 atm. at 600° C., and also for the blades of steam turbines. It was found, however, that this alloy was subject to stress-corrosion failures, and Chevenard introduced modifications which completely overcame this difficulty. He also

showed why these modifications were successful.

Another application of the work of our Honorary Member lay in the field of horology, where he introduced a variation of Elinvar for use in the balance-wheel springs of chronometers.

In addition to his research work, Chevenard took an active part in teaching. He was Professor at the Ecole des Mines at Saint-Etienne, where he had been a student, from 1919 to 1935, and also at the Ecole supérieure de Fonderie from 1923 to 1939. At present he is conducting a metallography course at the Ecole des Mines de Paris. Furthermore, he instituted courses in the works of his firm for apprentices, workmen, &c.

M. Chevenard has been the recipient of many honours. On three occasions

he was awarded a prize of the Académie des Sciences (1924, 1935, and 1943). He is lauréat of the Société des Ingénieurs civils de France; of the Société d'Encouragement; of the Société française de Navigation aérienne; and of the Centre national de la Recherche scientifique. He is a Past-President of the Société française de Physique and of the Association technique de Fonderie, and a Vice-President of the Société française de Métallurgie and of the Société de l'Industrie Minérale. In 1946 he was elected member of the Académie des Sciences, and this year he was made a doctor *honoris causa* of the University of Liège.

During the ceremonies connected with his induction into the Académie des Sciences one of the speakers pointed out that, as President of the French Physical Society, he emphasized the need for physicists to consider problems of value to industry, while as President of the Foundry Association he underlined the need for giving a scientific basis to the work of the Association.

Referring to his personal qualities another speaker said, on



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behalf of his old students, "... I also wish to express to you our affection, our very deep affection; your apparent shyness, caused by your modesty, hides a heart of gold which knows how to share the joys and sorrows of all those with whom you are associated". All M. Chevenard's friends will agree with the absolute truth of this testimony.

One final reference to the speakers at the Académie des Sciences celebrations. M. Chaudron told how M. Chevenard, when a school boy, was given as a subject for an essay "Describe how you would picture heaven", and the essay was a description of an enormous laboratory where one could operate beautiful electrical machines continuously and could prepare vast quantities of oxygen and hydrogen. Truly the child is father to the man.

J. L. H.

Professor Sir Geoffrey Taylor

Professor Sir Geoffrey Ingram Taylor, who has been elected an Honorary Member of the Institute of Metals, was born in 1886. On leaving University College School, he went to Trinity College, Cambridge, in 1905, and graduated in 1908. He was elected a Fellow of Trinity College in 1910.

From 1912 to 1914 he was Schuster Reader in Dynamical Meteorology, and in 1913 spent seven months on the whaling ship *Scotia* studying—by means of flying kites carrying instruments—the way heat is transferred from the atmosphere to the sea. This led to a study of the turbulent motion of fluids, which is still developing. At the outbreak of the 1914 war, being told by the War Office that meteorologists were useless in warfare, he was sent to the Royal Aircraft Factory at Farnborough to join the Aerodynamics Department. In 1915, however, finding it difficult to study this subject realistically from the ground, he joined the Royal Flying Corps and was granted his pilot's certificate and commissioned in June 1915.



After the war, he spent the summer of 1919 in Newfoundland, with the party which was competing for the prize for the first flight across the Atlantic. In 1919 he was elected a Fellow of the Royal Society, and later returned to Cambridge as lecturer in mathematics.

In 1922, Dr. Taylor became interested in metals and in 1923 gave, with Miss Elam (now Mrs. Tipper), the Bakerian Lecture to the Royal Society on "The Distortion of an Aluminium Crystal during a Tensile Test". This was followed by a number of

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other papers on the plastic properties of single crystals and of crystal aggregates. He published in 1934 the dislocation theory of strain hardening, and in 1938 delivered the Twenty-Eighth May Lecture to the Institute of Metals on "Plastic Strain in Metals". He was married in 1925.

During the 1939-1945 war, Professor Taylor was engaged in work on the dynamics of explosions. He was a member of the Civil Defence Research Committee and of the Ministry of Supply Advisory Council. Near the end of the war he was in New Mexico, where he was one of the team which made the atomic bomb and was present when the first one was tested. He was knighted, for his services, in 1944. Since the war, he has again become interested in the plasticity of metals.

Apart from his scientific work, Sir Geoffrey's main interest is sailing. He has cruised with his wife and a friend to the Lofoten Islands, 100 miles north of the Arctic Circle, and as far as Finland.

He received the Royal Medal of the Royal Society in 1933, the Copley Medal in 1944, and the (U.S.) Medal for Merit in 1947. He has been Yarrow Research Professor of the Royal Society since 1923. He gave the children's Christmas lectures at the Royal Institution in 1936 on "Ships".

Sir Geoffrey has received honorary doctorates of Oxford, Liverpool, London, British Columbia, Aachen, and Oslo, and Honorary Membership of the National Academy of Sciences (Washington), the Académie des Sciences (Paris), and the Koninklijke Nederlandsche Akademie van Wetenschappen (Amsterdam), and is also an Honorary Member of the Institution of Mechanical Engineers.

PERSONAL NOTES

MR. K. V. AIYER, B.A., of the Aluminium Company of Canada, has left Shawinigan Falls and is now at the firm's Etobicoke Works, Toronto, Canada.

MR. H. B. BARNARD has been re-elected a Member of Council of the National Association of Scrap Metal Merchants.

MR. H. N. BATES, M.I.Mech.E., has retired from his position as Managing Director of Specialloid, Ltd., London.

MR. JOHN CARTLAND, M.C., M.Sc., F.I.M., has been elected a Member of Council of the Printing and Allied Trades Research Association, and has also been appointed a member of the Research Committee.

MR. HORACE W. CLARKE, F.I.M., has been re-elected President of the British Non-Ferrous Metals Federation.

MR. J. DEARDEN, M.Sc., A.M.I.Mech.E., was appointed Acting Chief Metallurgist to the London, Midland and Scottish Railway, in succession to Dr. H. O'Neill, who has taken the Chair of Metallurgy at University College, Swansea.

MR. W. E. DUCKWORTH has left the English Electric Company, Ltd., Stafford, to take up an appointment as Metallurgist with British Insulated Callender's Cables, Ltd., Prescott, Lancs.

DR. C. A. EDWARDS, F.R.S., has been awarded the honorary degree of LL.D. of the University of Wales.

NEWS AND ANNOUNCEMENTS

MR. V. C. FAULKNER has been re-elected a Member of Council of the British Cast-Iron Research Association.

MR. V. Z. DE FERRANTI, M.C., Chairman and Managing Director of Ferranti, Ltd., Hollinwood, Lancs., received the honour of Knighthood in the New Year's Honours List.

MR. R. H. GARNER, B.Sc.Eng., Head of the Engineering and Science Department of the Blackpool Technical College, has been appointed Principal of a new technical college to be opened in the Burnbank industrial area of Glasgow, and expects to take up this new post in February.

MR. M. K. HALDAR, M.Sc., has left England. His address is The Indian Association for the Cultivation of Science, 210 Bowbazar Street, Calcutta 12, India.

DR. W. A. HAMOR, M.A., has received the honorary degree of Doctor of Science of the University of Louisville.

PROFESSOR D. HANSON, D.Sc., has been re-elected a Member of Council of the British Cast-Iron Research Association.

MR. P. W. HYDE, B.Sc., M.Met., L.I.M., has joined the staff of the Research Laboratories of The General Electric Company, Ltd., Wembley, Middlesex.

MR. KONRAD KORNFELD, Ing.Met., is now at the Polish Merchant Navy College, Lilford Hall Camp, Oundle, Northants.

DR. A. B. MCINTOSH, B.Sc., A.R.T.C., F.R.I.C., F.I.M., has left the Royal Technical College, Glasgow, to be in charge of the Metallurgical Department, Department of Atomic Energy, Risley, near Warrington, Lancs.

MR. T. F. PEARSON, M.Sc., F.I.M., has resigned his position with Colvilles, Ltd., and is now with the Consett Iron Company, Ltd., Consett, Co. Durham.

MRS. M. K. PETCH, B.A., was recently married and has left for Australia. She is now Mrs. M. K. McQuillan.

MR. E. PLAYER has been re-elected a Vice-President of the British Cast-Iron Research Association.

MR. S. RAMAMURTHY, B.Sc., has resigned his position as Research Metallurgist with Messrs. Jyoti, Ltd., Baroda. His present address is c/o S. Balakrishnan, 4 Normal School Quarters, Sitabuldi, Nagpur, India.

MR. W. S. ROBINSON has resigned from his position as Managing Director of the Zinc Corporation, Ltd., and also his seat on the Board. He has agreed to accept office as President of the Corporation.

DR. CYRIL S. SMITH has been appointed a member of the Metallurgical Advisory Committee to the (U.S.) National Bureau of Standards.

MR. E. J. D. SMITH, L.I.M., has left the Motor Industry Research Association to join the Refractories Section of the British Iron and Steel Research Association.

MR. W. A. SMITH, B.Sc., A.R.T.C., has left the General Manager's Department of Messrs. Stewarts and Lloyds, Ltd., Corby, and is now at the Company's Department of Research and Technical Development, Clyde Tube Works, Coatbridge, Lanarkshire.

NEWS AND ANNOUNCEMENTS

MR. H. E. N. STONE has left The Mond Nickel Company, Ltd., Birmingham, and is now with the staff of the British Iron and Steel Research Association at the National Physical Laboratory, Teddington, Middlesex.

DR. T. A. TAYLOR, M.Sc., A.R.I.C., A.I.M., has left Power Jets, Ltd., Lutterworth, to take up an appointment at the National Gas Turbine Establishment, Whetstone, Leicester.

MR. J. S. THOMPSON, B.Sc., L.I.M., has taken up the post of Assistant Works Manager to the Newcastle-upon-Tyne Zinc Oxide Company, Ltd., Birtley, Co. Durham.

MR. GEOFFREY W. TURNER, L.I.M., has been appointed Technical Assistant, Wire Mills, at the Elliott Works of the Metals Division, Imperial Chemical Industries, Ltd., Birmingham. He has also recently been awarded an Associateship of the Central Technical College, Birmingham.

DEATH

The Editor regrets to announce the death on 20 November 1947, of Mr. Norman Arbuthnot Clark, B.Sc. (Eng.), Management Consultant, of London; formerly Director and General Manager of Darwins, Ltd., of Sheffield.

Note : Will members (in addition to informing the Institute's administrative department of changes of address, occupation, &c.) kindly notify the Editor, separately, of all changes of occupation, appointments, awards of honours and degrees, &c., as these are matters which interest their fellow members. Such notes should reach the Editor not later than the 21st day of each month, for publication in the next month's issue.

ELECTION OF ORDINARY MEMBERS AND STUDENT MEMBERS

The undermentioned 29 Ordinary Members and 38 Student Members were elected on 30 December 1947.

As Ordinary Members

ASHWORTH, Fred, B.Sc., Research Physicist, Research Department, Metropolitan-Vickers Electrical Company, Ltd., Trafford Park, Manchester 17.

BHAT, Professor Uppinangady Vaman, B.Sc. (Met.), Assistant Professor of Metallurgy, Benares Hindu University, India.

BLOOMER, Keith Malcolm, Metallurgical Analyst, Allcock and Company (Metals), Ltd., Hockley, Birmingham.

DEARDEN, John, M.Sc., Acting Chief Metallurgist, Scientific Research Department, London, Midland and Scottish Railway Company, London Road, Derby.

DOCKER, Sir Bernard Dudley Frank, Chairman, James Booth and Company, Ltd., 69 Victoria Street, London, S.W.1.

DUWEZ, Professor Pol, D.Sc., Associate Professor of Mechanical Engineering, California Institute of Technology, Pasadena, Cal., U.S.A.

NEWS AND ANNOUNCEMENTS

- ERSKINE, John, Representative, Park and Paterson, Ltd., Glasgow, E.1.
- DE FERRANTI, Vincent Ziani, M.C., Chairman and Managing Director, Ferranti, Ltd., Hollinwood, Lancashire.
- GILBY, Leonard Charles, Chief Metallurgist, Associated Equipment Company, Ltd., Southall, Middlesex.
- GOODEVE, Sir Charles Frederick, D.Sc., F.R.S., Director, British Iron and Steel Research Association, 11 Park Lane, London, W.1.
- GORDON, Henry James, Managing Director, C. F. Sachs and Company, Ltd., 27/35 Stockwell Place, Glasgow, C.1.
- HADDEN, Stanley E., Research Assistant, British Non-Ferrous Metals Research Association, Euston Street, London, N.W.1.
- HUNTER, Philip Vassar, C.B.E., Chairman, Thomas Bolton and Sons, Ltd., Norfolk House, Norfolk Street, London, W.C.2.
- KIRKWOOD, Robert Smith, Director, Vincent Engineering Company, Ltd., and Scottish Industrial Products, Ltd., 26 Blythswood Square, Glasgow, C.2.
- LOW, Professor John R., Jr., B.S., D.Sc., Professor of Metallurgy, Pennsylvania State College, State College, Pa., U.S.A.
- McKERROW, George, M.A., Director, Cutanit, Ltd., and Compound Electro Metals, Ltd., Glencaird, Bargrennan, Newton Stewart, Scotland.
- MAZZOLENI, Professor Francesco, Professore di Technologie Speciali nella Facoltà d' Ingegneria dell' Università di Napoli, via Mezzocannone 16, Napoli, Italy.
- MONVOISIN, Jean, Ingénieur Chimiste, Chef du Laboratoire de Recherches Métallurgiques, Etablissements J. J. Carnaud et Forges de Basse Indre, 71 avenue Eduard Vaillant, Billancourt (Seine), France.
- MOYLE, Clifford Edwin, B.Sc. (Eng.), Victor Moyle and Company, Founders and Pattern Makers, 38 Park Road, Hampton Wick, near Kingston-on-Thames, Surrey.
- MULLER, Albert, M.E.E., Ph.D., Metallurgical Engineer, Air Reduction Sales Company, Inc., Murray Hill, N.J., U.S.A.
- NIX, Foster C., Ph.D., Director of Research, The Sharples Corporation Research Laboratories, 201 Spring Garden Street, Philadelphia 23, Pa., U.S.A.
- PATERSON, James, Director, Park and Paterson, Ltd., Glasgow, E.1.
- PATERSON, William, J.P., Director and Chairman, Park and Paterson, Ltd., 6 Buchanan Gardens, Mount Vernon, Glasgow, E.2.
- PRAIN, Ronald Lindsay, O.B.E., Director, Roan Antelope Copper Mines, Ltd., Selection Trust Buildings, Masons Avenue, London, E.C.2.
- REIM, Ernest Philip, B.Sc. (Eng.), Senior Lecturer in Mechanical Engineering, Natal University College, P.O. Box 1525, Durban, South Africa.
- RIDGEWELL, Percy, Head of Technical Information Section, The Mullard Radio Valve Company, Ltd., Cross Oak Lane, Salfords, near Redhill, Surrey.
- RYSKA, Charles, Dr. Ing., D.Sc., Research Engineer, Engine Works, Ceskomoravská-Kolben-Daněk, Prague, Czechoslovakia.
- SUESS, Franz Joseph, Managing Director, Metal Supplies, Ltd., 27 Chancery Lane, London, W.C.2.

NEWS AND ANNOUNCEMENTS

WIRDNAM, Geoffrey Guy, Works Manager, H. Roller and Company, Ltd., 32/36 Rosebery Avenue, London, E.C.1.

As Student Members

- AKINCI, Abdulmecit, Student of Metallurgy, Birmingham University.
- ARTHUR, Patrick Terence, B.Sc. (Hons. Met.), Research Assistant, Department of Metallurgy, Royal Technical College, Glasgow.
- BAILEY, Edwin Anthony, Student of Metallurgy, Royal School of Mines, Prince Consort Road, London, S.W.7.
- BALE, Ernest Samuel, Senior Laboratory Assistant, British Non-Ferrous Metals Research Association, Euston Street, London, N.W.1.
- BRACEWELL, Richard Henry, Assistant Chemist and Metallurgist, Sterling Metals, Ltd., Northey Road, Foleshill, Coventry.
- BUTLER, Ronald Arthur, Student of Metallurgy, Liverpool University.
- CARTER, John James, Student of Metallurgy, The University, Melbourne, Australia.
- CONDÉ, John Frederick George, B.Sc. (Hons.), Research Metallurgist, The Mond Nickel Company, Ltd., Development and Research Department, Wiggan Street, Birmingham, 16.
- COULSON, Kenneth John, Student of Metallurgy, Birmingham University.
- DAVIES, Morgan Harris, B.Sc. (Hons.), Research Metallurgist, Associated Electrical Industries Research Laboratories, Aldermaston Court, Aldermaston, Berkshire.
- DEBIE, Rafael, Student of Metallurgy, Royal School of Mines, London, S.W.7.
- EDELEANU, Constantin, B.A., Research Student, Corrosion Laboratory, Metallurgy Department, Cambridge University.
- FIGGIS, David Lenox, Student of Metallurgy, Liverpool University.
- HEISCH, Godfrey Edward, B.Sc., A.R.S.M., Student of Metallurgy, Royal School of Mines, Prince Consort Road, London, S.W.7.
- HOGGART, John Stewart, Student of Metallurgy, The University, Melbourne, Australia.
- HOMER, Frank, Chemical Laboratory Assistant, J. A. Crabtree and Company, Ltd., Lincoln Works, Walsall, Staffordshire.
- JENKINS, Alexander Elliott, Student of Metallurgy, The University, Melbourne, Australia.
- MARLES, Douglas, Chief Metallographer, British Cast-Iron Research Association, Alvechurch, Birmingham.
- MIODOWNIK, Alfred Peter, Student of Metallurgy, Battersea Polytechnic, London, N.W.11.
- NELLER, Robert Frank, Technical Assistant, Metal Manufacturers, Ltd., 18 Strathearn Avenue, Wollongong, N.S.W., Australia.
- OTTE, Henry Martin, Student, Birmingham University.
- PARRY, Paul I., Student, Liverpool University.
- PATERSON, Mervyn Silas, B.E., Research Student (Metal Physics), Clare College, Cambridge.
- PICK, John Julius, Student of Metallurgy, Liverpool University.
- POWELL, John Robert, B.Sc., Student of Metallurgy, Nuffield Central Research Laboratory, Courthouse Green, Coventry.
- POYNER, Martin Derick Marsh, Laboratory Assistant, J. A. Crabtree, Ltd., Lincoln Works, Walsall, Staffordshire.

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- PRICE, Peter, Student, Physics Department, University College of North Wales, Bangor.
- PROFFITT, Henry Joseph, Research Worker, High-Duty Alloys, Ltd., Slough, Buckinghamshire.
- ROGERS, Colin Francis, Metallurgical Assistant, United Wire Works (Birmingham), Ltd., Adderley Street, Birmingham, 9.
- SHAW, John Arthur, Student, Leeds University.
- SMYTHE, Samuel Brian, Student of Metallurgy, Liverpool University.
- SRAWLEY, John Ewart, Student of Metallurgy, Birmingham University.
- STREET, Brian Godfrey, Student of Metallurgy, Liverpool University.
- TURNBULL, George Alexander, B.Sc., Research Metallurgist, Research Laboratories, The British Aluminium Company, Ltd., Kinlochleven, Argyllshire.
- TYLDESLEY, William R., Student of Metallurgy, Manchester University.
- WARING, James, Student of Metallurgy, College of Technology, Manchester.
- WATKINSON, John Francis, Student of Metallurgy, Sheffield University.
- WILLAN, Ronald Robert, Student of Metallurgy, Liverpool University.
- WILLMOTT, Peter Lloyd, Chemist, Stewarts and Lloyds, Ltd., Corby, Northamptonshire.

LOCAL SECTIONS NEWS

MEMBERS' CHANGES OF ADDRESS

When members, by changing their addresses, pass into the area of another Local Section or Associated Society they are particularly requested to notify the Honorary Secretary of the Local Section in that area.

The addresses of the Honorary Secretaries of the Local Sections and Associated Societies are :

Birmingham : Mr. E. H. Bucknall, M.Sc., 53 Halesowen Road, Quinton, Birmingham 32.

London : Dr. E. C. Rhodes, c/o The Mond Nickel Company, Ltd., Development and Research Department, Bashley Road, London, N.W.10.

Scottish : Mr. Matthew Hay, 132 Elliot Street, Glasgow, C.3.

Sheffield : Dr. W. R. Maddocks, Department of Metallurgy, The University, St. George's Square, Sheffield.

South Wales : Mr. D. W. Hopkins, B.Sc., Metallurgical Department, University College, Singleton Park, Swansea.

Leeds Metallurgical Society : Mr. W. J. G. Cosgrave, 24 Well-house Road, Leeds 8.

Manchester Metallurgical Society : Mr. J. A. Todd, B.Sc., Imperial Chemical Industries, Ltd., Metals Division, Broughton Copper Works, Manchester.

OTHER NEWS

FRENCH METALLURGICAL SYMPOSIA

From 6 to 15 October 1947 three symposia were held in Paris, attended by many eminent French and foreign metallurgists, metallographers, foundrymen, chemists, and engineers.

First of all there was the "Journées de la Corrosion", organized under the auspices of the "Commission des États de Surface", under the Chairmanship of Professor A. Portevin. Professor G. Chaudron was the indefatigable organizer.

Afterwards there was the "Congrès de la Société française de Métallurgie", presided over by M. A. Aron, Vice-President of the "Chambre syndicale de la Sidérurgie". Many papers were presented and M. Aron was assisted by Professor A. Portevin (who was Chairman in 1946) and Professor P. Chevenard (who will be Chairman in 1948), as it was necessary to hold simultaneous sessions.

These two meetings were followed by the annual congress of the Association Technique de Fonderie, with Professor P. Bastien as Chairman.

Among those present were: Drs. U. R. Evans, C. H. Desch, W. H. J. Vernon, R. Seligman, N. Stuart, Gulbransen, and Pray, Professors Carl A. F. Benedicks, Jenicek, Thyssen, G. Batta, and F. Meunier, and Mr. F. Hudson.

The papers made important contributions to metallurgical problems. In the field of corrosion, in particular, a coherent theory seemed to be established on the electrochemical mechanism of wet corrosion, and notable progress was made in the understanding of dry corrosion.

Important contributions came from the investigations of the English school, under Dr. U. R. Evans, and of the French school, working under Professor G. Chaudron.

SYMPOSIUM ON THE FAILURE OF METALS BY FATIGUE

The publication of the proceedings of this symposium, which was held in Melbourne in December 1946, has been delayed by difficulties in the printing industry. The volume is not likely to be ready for despatch to subscribers until March 1948.

HIGH VACUUM SYMPOSIUM

Leading scientists from the United States and abroad gathered on 30 October 1947 in Cambridge, Mass., U.S.A., to attend a High Vacuum Symposium, sponsored by the National Research Corporation of Cambridge, Mass., in co-operation with the Division of Industrial and Engineering Chemistry of the American Chemical Society. The meeting was devoted to discussions of the latest techniques and developments in the rapidly growing field of high vacuum.

Under the general Chairmanship of Mr. Richard S. Morse, President of the National Research Corporation, assisted by Dr. John R. Bowman of the Mellon Institute of Industrial Research, the various sections of the meeting stressed the different aspects

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of high vacuum. Dr. Saul Dushman, Assistant Director of the General Electric Research Laboratories, led the group discussing the early history of high vacuum and basic research which led to its present day development. Dr. Bowman's session covered the field of high vacuum distillation which has led to the production in commercial quantities of vitamins, plasticizers, and many other substances which could not be fractionated previous to the development of high vacuum. Problems in connection with large-scale commercial installation of high vacuum equipment were discussed under the leadership of Dr. John C. Hecker, Works Manager of Distillation Products, Inc., and formerly production manager of one of the atomic energy production plants at Oak Ridge, Tenn.

The many new developments in the field of high vacuum metallurgy were discussed by scientists led by Dr. John Chipman, Head of the Metallurgical Department of the Massachusetts Institute of Technology. These discussions covered recent developments in the commercial production of magnesium, gas-free copper and improvements in sintering, as well as vacuum heat-treating.

A. L. Schroeder, Director, Dehydration Department, National Research Corporation, was Chairman of the closing session which discussed the various engineering problems arising in the field of vacuum dehydration.

COPPER AND ITS ALLOYS AND ZINC; GLOSSARY OF TERMS

The British Standards Institution has recently issued a "Glossary of Terms Applicable to Wrought Products in Copper, Zinc, Brass, and Other Copper Alloys" (obtainable, price 2s., post free, from the Institution, 28 Victoria Street, London, S.W.1), which should have a wide interest. This British Standard has been approved by the Non-Ferrous Metals Industry Standards Committee and endorsed by the Chairman of the Engineering Divisional Council.

The request for this glossary came from the B.S.I. Committee preparing standards for copper for electrical purposes. The matter was referred to the British Non-Ferrous Metals Federation, which set up a committee representative of the trade associations concerned; that committee had before it definitions published by the American Copper and Brass Research Association.

The definitions have been related to current commercial practice and due consideration has been given to H.M. Customs definitions.

While the scope of this glossary has been limited to wrought products, it is hoped that further sections will be published covering raw materials, cast shapes, and possibly certain fabricated products.

RESEARCH: A JOURNAL OF SCIENCE AND ITS APPLICATIONS

This new monthly periodical has recently been published by Butterworth Scientific Publications, Bell Yard, Temple Bar, London, W.C.2; the price per copy is 3s. 6d., and the annual subscription is 45s., post free.

The journal is edited by Dr. P. Rosbaud and Mr. M. D. Rexworthy, with the assistance of a strong editorial board under the Chairmanship of Sir John Anderson, with Professor R. S. Hutton as Secretary.

The object of *Research* is to forge a link between fundamental science and its applications in industry and elsewhere. The first issues have reached a high standard, and this journal should fill a gap in the scientific literature.

DIARY FOR FEBRUARY

LOCAL SECTIONS MEETINGS

Birmingham Local Section.—Dr. W. Steven: "Modern Developments in Hardenability Testing". (James Watt Memorial Institute, Great Charles St., Birmingham, Thursday, 5 February, at 6.30 p.m.)

Birmingham Local Section.—Mr. C. A. H. Jahn: "The Metallurgy of Gold, Silver, and Platinum". (James Watt Memorial Institute, Great Charles St., Birmingham, Thursday, 26 February, at 6.30 p.m.)

London Local Section.—Dr. H. Sutton: "Metallurgical Problems of Importance in Aircraft". (4 Grosvenor Gardens, London, S.W.1, Thursday, 12 February, at 7 p.m.)

Scottish Local Section.—Dr. L. B. Pfeil: "Gas Turbine Materials". (Institution of Engineers and Shipbuilders in Scotland, 39 Elmbank Crescent, Glasgow, C.3, Monday, 9 February, at 6.30 p.m.)

Sheffield Local Section.—Lecture on "Powder Metallurgy". Mining Lecture Theatre, Department of Applied Science, The University, St. George's Sq., Sheffield, Friday, 20 February, at 7.30 p.m.)

South Wales Local Section.—Mr. W. H. Tait: "Bearings". (Y.M.C.A., Swansea, Tuesday, 10 February, at 6.30 p.m.)

OTHER MEETINGS

MONDAY, 2 FEBRUARY

Cleveland Institution of Engineers.—R. P. Towndrow: "Some Aspects of Modern Blast-Furnace Operation". (Cleveland Scientific and Technical Institution, Corporation Rd., Middlesbrough, at 6.30 p.m.)

TUESDAY, 3 FEBRUARY

Electrodepositors' Technical Society, Midlands Centre.—Dr. J. W. Cuthbertson: "The Electrodeposition of Speculum"; W. H. Sawyer: "Practical Aspects of Speculum Plating". (James Watt Memorial Institute, Great Charles St., Birmingham 3, at 6.30 p.m.)

Institute of the Motor Industry, Workington Centre.—Mr. Ellender: "Oxy-Acetylene Welding for the Use of Vehicle Maintenance in all Aspects". (Technical School, Workington, at 7.30 p.m.)

WEDNESDAY, 4 FEBRUARY

Birmingham University Metallurgical Society.—Visit to Round Oak Steel Works, Ltd., Brierley Hill, Staffs.

Institution of Works Managers.—G. T. Harwood-Stamper:

NEWS AND ANNOUNCEMENTS

"Factory Management". (Waldorf Hotel, Aldwych, London, W.C.2, at 6.30 p.m.)

Royal Society of Arts.—L. Urwick, O.B.E., M.C., M.A.: "Education for Management". (The Society, John Adam St., Adelphi, London, W.C.2, at 2.30 p.m.)

THURSDAY, 5 FEBRUARY

Birmingham University Metallurgical Society.—Presidential Address. (University Buildings, Edgbaston, Birmingham, at 5 p.m.)

Institute of Welding.—R. G. Braithwaite: "Some Problems of Approach to Welding Design". (Cleveland Scientific and Technical Institution, Corporation Rd., Middlesbrough, at 7.30 p.m.)

Institute of Welding, South London Branch.—H. A. Cadwell: "New Recommendations for the Design and Fabrication of Welded Structures". (Institute of Marine Engineers, 85-88 The Minories, London, E.C.3, at 6.30 p.m.)

Leeds Metallurgical Society.—C. P. Miller: "Investigation of Failures". Meeting with Leeds Association of Engineers. (Main Lecture Theatre, Chemistry Department, The University, Leeds, at 7 p.m.)

WEDNESDAY, 11 FEBRUARY

Geological Society of London.—Ordinary Evening Meeting. (Burlington House, Piccadilly, London, W.1, at 5 p.m.)

Manchester Metallurgical Society.—T. F. Russell: "Isothermal Heat-Treatment of Alloy Steels". (Engineers' Club, Albert Sq., Manchester, at 6.30 p.m.)

MONDAY, 16 FEBRUARY

Electrodepositors' Technical Society.—P. Spiro, B.Sc.: "The Production of Electro-Formed Moulds for Plastics and Die-Castings". (Northampton Polytechnic, St. John St., Clerkenwell, London, E.C.1, at 5.30 p.m.)

Sheffield Society of Engineers and Metallurgists.—Exhibition of technical films. (Royal Victoria Station Hotel, Sheffield, at 6.15 p.m.)

TUESDAY, 17 FEBRUARY

Institute of the Motor Industry, Gloucester Centre.—R. J. Brown: "Failures of Materials in Service". (The Belle Vue Hotel, The Strand, Cheltenham, at 7.30 p.m.)

WEDNESDAY, 18 FEBRUARY

Birmingham University Metallurgical Society.—Visit to E. J. and J. Pearson, Ltd. (Refractories), Brierley Hill, Staffs.

Institute of the Motor Industry, York Centre.—P. P. Love: "Bearing Failure—Analysis and Prevention". (Rechabite Hall, Clifford St., York, at 7.30 p.m.)

NEWS AND ANNOUNCEMENTS

Institution of Structural Engineers, London Graduates' and Students' Section.—O. Bondy: "Modern Trends in Structural Welding". (11 Upper Belgrave St., London, S.W.1, at 6 p.m.)

THURSDAY, 19 FEBRUARY

Institute of Welding, South London Branch.—Speakers and discussion on "Practical Methods of Welding Cast-Iron". (County Technical College, Guildford, at 6.30 p.m.)

Institution of Mining and Metallurgy.—General Meeting. (Geological Society of London, Burlington House, Piccadilly, London, W.1, at 5 p.m.)

SATURDAY, 21 FEBRUARY

Swansea and District Metallurgical Society.—W. F. Cartwright: "Engineering Aspects of a Modern Steel Plant". (Royal Institution, Swansea, at 6.30 p.m.)

MONDAY, 23 FEBRUARY

Institution of Works Managers.—J. A. Gunn: "The Bearing of Sales on Works Management". (Institution of Engineers and Shipbuilders in Scotland, 39 Elmbank Crescent, Glasgow, C.3, at 7 p.m.)

THURSDAY, 26 FEBRUARY

Birmingham University Metallurgical Society.—Dr. J. W. Jenkin: "The Functions of an Industrial Research Organization". (University Buildings, Edgbaston, Birmingham, at 5 p.m.)

APPOINTMENTS VACANT

To conform to the requirements of the Control of Engagements Order, 1947, these advertisements are published for the information of those only who are "excepted persons" under the Order.

CHIEF METALLURGIST required for large non-ferrous metal works in Midlands. The position carries senior executive responsibility and status and offers considerable scope. Salary will be in keeping with the position. Candidates should possess an honours degree in metallurgy, be aged between 30 and 40, and should have had practical experience in the non-ferrous metals industry. Apply, giving full details of age and experience, to Box No. 222, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGIST required to take charge of mechanized foundry in Lancashire area producing brass and gun-metal castings for pressure work. Good opening for young man with personality and drive who can control labour with efficiency and obtain maximum output. Please state age, experience, qualifications, references, and salary required. Reply to Box No. 221, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGY OR PHYSICAL CHEMISTRY GRADUATE, preferably with some research experience, is required for metallurgical research work in connection with high-temperature materials. Apply by letter, stating age and qualifications, to the Director, Research Laboratories of The General Electric Co., Ltd., North Wembley, Middlesex.

SENIOR METALLURGIST, aged 30-45, with degree or equivalent qualification and industrial experience, required by long established heavy non-ferrous foundry in the North-West. Salary according to qualifications but approximately £1000 p.a. Reply to Box No. 223, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

HOT-SHORTNESS OF SOME HIGH-PURITY 1119 ALLOYS IN THE SYSTEMS ALUMINIUM- COPPER-SILICON AND ALUMINIUM-MAG- NESIUM-SILICON.*

By P. H. JENNINGS,† Ph.D., B.Sc., STUDENT MEMBER, A. R. E. SINGER,‡
Ph.D., B.Sc., MEMBER, and W. I. PUMPHREY,§ M.Sc., MEMBER.

(From the School of Industrial Metallurgy, University of Birmingham.)

SYNOPSIS.

The methods of investigating hot-shortness previously used for aluminium-silicon and aluminium-iron-silicon alloys, namely the carrying out of a ring-casting test and a restrained-weld test, have been applied to the systems aluminium-copper-silicon and aluminium-magnesium-silicon. The results have enabled ternary hot-shortness diagrams to be constructed for both systems, from which a knowledge can be gained of the relative welding and casting behaviour of alloys in the ranges copper 0-10, silicon 0-4%; and magnesium 0-10, silicon 0-5%.

A study has also been made of the tensile properties of some of the aluminium-copper-silicon alloys possessing relative freedom from hot-shortness.

The reasons for hot-shortness in the alloys investigated seem to be constitutional, as was found previously to be true for binary aluminium-silicon alloys. The original hot-shortness theory has been extended so as to relate to ternary systems, but its application is less straightforward, particularly in aluminium-magnesium-silicon alloys, in which the presence of the Mg_2Si phase is a complicating factor.

I.—INTRODUCTION.

THE present work was carried out on similar lines to a previous investigation of the hot-shortness of aluminium-iron-silicon alloys,¹ but it differs in one important respect. Whereas the effects of small additions of silicon and iron were studied with the object of fixing the maximum content of impurities allowable in commercial aluminium, the present research deals with a more extensive range of alloys. The results cover the range copper 0-10 and silicon 0-4% in the aluminium-copper-silicon system, and magnesium 0-10 and silicon 0-5% in the aluminium-magnesium-silicon system.

Several industrial alloys lie within the ternary fields investigated, so that the results have a direct practical application. Alloys containing approximately 3% copper and 4% silicon, for instance, have good

* Manuscript received 29 April 1947.

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casting properties and have long been used in the foundry. More recently, similar alloys have been prepared in the wrought form, and it has been found that they can be welded into complex structures without danger of cracking. The alloys of this type so far used in industry have been confined to limited areas of the aluminium-copper-silicon ternary diagram, and the present work enables some estimate to be made of the suitability of other alloys for casting and welding.

In the aluminium-magnesium-silicon system, there are two main types of industrial alloys in relation to which the results of the present work are likely to be of interest. The first type includes the alloys of aluminium with 3-7% magnesium, which sometimes weld unsatisfactorily on account of gas porosity. To remedy this tendency to porosity, silicon is sometimes added to the extent of 0.5-1.0%. The likely effect of this addition on the hot-shortness and tendency to weld cracking can be deduced from the present results. The second type comprises the heat-treatable alloys which depend for their properties on the precipitation of magnesium silicide. These alloys are particularly prone to hot cracking in both castings and welds, and the possibility exists of minimizing this defect by slight alterations in composition.

II.—PREVIOUS WORK.

The literature contains several references to the hot-shortness of individual alloys of the aluminium-copper-silicon and the aluminium-magnesium-silicon type, but no comprehensive work on either ternary system appears to have been done.

Dix and Lyon² found that alloys containing 3-5% copper with 3-5% silicon were slightly less prone to hot-short cracking and other casting defects than the more generally used 8%-copper alloy, though the mechanical properties were similar. Tichy³ found that the poor casting properties of an alloy of the Duralumin type, obtained by remelting aircraft scrap, could be improved by additions of silicon, and an alloy containing principally copper 3 and silicon 5% was developed which had good mechanical properties and was suitable for sand- or die-casting.

There are several references to the poor welding properties of heat-treatable aluminium-magnesium-silicon alloys, particularly their tendency towards cracking.^{4, 5} This defect is frequently overcome by the use of a 5%-silicon filler rod—an unsatisfactory procedure since heat-treatment of the welded structure is then less effective.

More work has been done on the hot-shortness of alloys in the binary systems of aluminium with silicon, copper, and magnesium. In experi-

ments with a series of ten aluminium-silicon alloys containing from 0.07 to 12.2% silicon, Verö ⁶ found by means of a casting test that there was a gradual increase in hot-shortness with increasing silicon content up to 1.6%, followed by a sudden decrease, and that alloys containing 1.88% silicon or more were not hot-short.

Bochvar and Makimdzhanova ⁷ devised a casting test which they used on a very limited number of aluminium-silicon and aluminium-copper alloys, and found that the susceptibility to cracking decreased as the alloy content was raised. The effect was most marked between 0.2 and 4% silicon and 4 and 8% copper. The same tendency was observed by Lees.⁸

Ridder ⁹ carried out restrained-weld tests on several aluminium alloys, but observed no tendency towards the formation of fissures in aluminium-magnesium alloys. He stated, however, that the higher the magnesium content the greater is the likelihood of cracking in such alloys. Marshall,¹⁰ on the other hand, found the maximum tendency to weld cracking at 1% magnesium.

III.—EXPERIMENTAL WORK.

The methods used in the present work for investigating hot-shortness were the same as those employed in earlier work on aluminium-silicon alloys of commercial purity ¹¹ and aluminium-iron-silicon alloys,¹ namely a ring-casting test and a restrained-weld test. In each of these tests, the stresses set up during cooling of the solidifying metal cause cracking in hot-short alloys.

In the aluminium-copper-silicon system, both ring-casting and restrained-weld experiments were carried out on aluminium-rich alloys covering the whole field in which appreciable hot-shortness was found, this field extending to 10% copper and 4% silicon. About 50 binary and ternary alloys were examined by the ring-casting test, whereas only 25 alloys were tested by restrained welding, owing to the necessity of producing sheet in each alloy. A few of the results for binary aluminium-silicon alloys were obtained during the experiments on the aluminium-iron-silicon system.¹ The ternary field was covered by investigating the effect of varying the percentage of one alloying element at a time. This method produces a rectangular pattern or "grid" when the alloy compositions are plotted on the ternary diagram.

The experiments with aluminium-magnesium-silicon alloys were restricted in the main to ring-casting tests, which were carried out on about 50 alloys besides the binary aluminium-silicon alloys for which the results were already known. A few restrained-weld tests were made

in order to determine whether the close agreement between the two tests already established in other alloy systems was maintained. Again, aluminium-rich alloys extending to the fringe of the hot-short composition range were investigated, i.e. up to 10% magnesium and 5% silicon. The ternary field was covered by six radial series of alloys, each beginning with aluminium and extending to the limit of the hot-short zone. One of the series was arranged to be coincident with the aluminium-magnesium silicide quasi-binary line, and in the other five the ratio of magnesium to silicon was 6 : 1, 3 : 1, 1 : 1, 1 : 2, and 1 : 4. The spacing of the alloys is shown in Fig. 3, and was closer near the aluminium corner of the diagram, where it was anticipated that the change of properties with composition would be most rapid.

This method of covering a ternary field has been used before by Kuznetsov and Makarov¹² when investigating the structure and ageing properties of aluminium-magnesium-silicon alloys.

1. Preparation of the Alloys.

The melting, casting, and rolling procedure was similar to that adopted in previous investigations.^{1, 11} For the ring-casting test, 400-g. charges were prepared, a quantity sufficient for four rings. The alloys for the welding test were cast as slabs measuring $6\frac{1}{2} \times 4 \times \frac{9}{16}$ in. The materials used were super-purity aluminium, two temper-alloys containing respectively 20% silicon and 33% copper, and 99.8%-purity magnesium. The temper alloys were prepared from super-purity aluminium, 98%-purity silicon, and electrolytic copper.

The aluminium was melted in a small gas-fired furnace, the crucible being lined with alumina cement to prevent contamination with silicon. The alloy additions were made under a suitable flux cover, and particular care was taken to prevent oxidation of magnesium by keeping the charge below 700° C. when adding this element.

The slabs for the production of the welding sheet were hot rolled at 400°–450° C. to 0.13 in. and cold rolled to the finishing thickness of 0.080 in. Annealing treatments at 400° C. followed both hot and cold rolling.

2. Ring-Casting Test.

The ring-casting test consists in casting a ring, of 2.3 in. external and 1.5 in. internal dia. and approximately 0.75 in. high, in an open cast-iron mould.

The pouring temperature is carefully controlled. In the present work, it was standardized at 100° C. above the liquidus, determined from the appropriate equilibrium diagram.^{13, 14} Radial cracks are

produced across the section of the ring in hot-short alloys during cooling, and the susceptibility of a particular alloy to cracking is assessed by measuring, at room temperature, the total length of cracking on all surfaces of the ring.

3. *Restrained-Weld Test.*

A 3-in. butt weld is made between two 14-gauge sheets clamped in a jig, by the left-ward method using an oxy-acetylene flame of a standardized size and a filler rod of the same material as the test sheets. The tensile stresses set up by contraction during cooling cause cracks to form in hot-short alloys, either at the centre or edges of the weld bead. The total length of these cracks on the top and bottom surfaces of the weld is measured.

For further information regarding the apparatus and methods used in the two tests, reference should be made to the original paper.¹¹

4. *Tensile Testing of Aluminium-Copper-Silicon Alloys.*

The work on the hot-shortness of aluminium-copper-silicon alloys was supplemented by tensile tests on several of the alloys showing good hot-shortness characteristics. These tests were done on both welded and un-welded material in the annealed and heat-treated conditions. The results are summarized in Appendix I.

IV.—EXPERIMENTAL RESULTS.

From 4 to 16 ring castings were made in each alloy, the number depending on the order of agreement obtained, and an average of the results was taken. Individual results seldom differed from the mean by more than $\pm 30\%$, and in many cases the agreement was much closer. As was found in earlier work and mentioned in a previous paper,¹¹ cracking in the rings was observed to take place at a temperature in the region of the solidus, subsequent cooling serving to increase the width but not the number or length of the cracks. When the rings were broken open, the cracks were seen to be intercrystalline and to show up clearly the junction plane between the outer and inner columnar crystals.

The number of restrained-weld tests carried out on each alloy varied from two to four (four being the more usual number), and the results, expressed in inches of cracking, were averaged. The presence of flux residues prevented detailed observations of the cracking process, but it was clear that formation of the cracks took place very soon after welding, and that, as in the ring-casting test, the subsequent contraction caused them to widen and become more easily distinguishable.

1. *Aluminium-Copper-Silicon Alloys.*(a) *Ring Castings.*

The individual experimental results for ring castings and restrained welds are given in Tables I and II respectively, from which the ternary diagrams of cracking, Figs. 1 and 2, have been constructed.

TABLE I.—*Cracking of Ring Castings in Aluminium-Copper-Silicon Alloys.*

(a).			(c).		
Nominal Composition.		Average Length of Cracking, in.	Nominal Composition.		Average Length of Cracking, in.
Cu, %.	Si, %.		Cu, %.	Si, %.	
0	0	0	0.25	0.25	7.1
0.1	0	1.4	0.5	0.5	7.4
0.2	0	4.5	1	1	7.2
0.35	0	6.2	1	2	4.0
0.5	0	7.3	1	3	0.3
1	0	6.9	1	4	0
1.5	0	6.8	2	1	5.7
2	0	5.8	2	2	1.9
2.5	0	6.2	2	3	0.6
3	0	6.7	2	4	0
3.5	0	6.8	3	1	5.9
4	0	5.6	3	2	1.7
5	0	2.8	3	3	0.4
7	0	1.3	4	1	4.8
9	0	0.7	4	2	1.0
(b).			4	3	0
0	0.1	5.6	4	12	0
0	0.25	6.9	5	1	3.0
0	0.6	6.9	5	2	0.9
0	1	5.4	5	3	0.2
0	1.5	4.0	5	12	0
0	2	2.3	6	1	2.9
0	4	0	6	2	0.3
0	8	0	7	1	1.9
0	12	0	7	2	0.6
0		0	8	1	1.0

In the two binary series, aluminium-silicon and aluminium-copper, the first alloying additions to aluminium cause a steep initial increase in hot-shortness, followed by a more gradual decrease to a very low value. In the aluminium-silicon series, the maximum hot-shortness occurs at about 0.4% silicon, while in the aluminium-copper series there appear to be two maxima, and all alloys containing between 0.5 and 4.0% copper are very hot-short. Alloy additions of more than 0.5% silicon and 4% copper cause a reduction in hot-shortness, which

reaches a negligible value at about 3% silicon and 10% copper. It seems, therefore, that although the hot-shortness characteristics are similar in each system, silicon is far more effective than copper in reducing cracking.

In the case of the ternary alloys, it can be seen that the addition of silicon to an aluminium-copper alloy and, more particularly, of copper

TABLE II.—*Cracking of Restrained Welds in Aluminium-Copper-Silicon Alloys.*

(a).			(b).			(c).		
Nominal Composition.		Average Length of Cracking, in.	Nominal Composition.		Average Length of Cracking, in.	Nominal Composition.		Average Length of Cracking, in.
Cu, %.	Si, %.		Cu, %.	Si, %.		Cu, %.	Si, %.	
0	0	0.05	0	0.1	1.0	0.5	0.75	8.3
0.25	0	1.2	0	0.25	3.7	1	1.5	5.0
1	0	5.6	0	0.4	5.4	2	1.5	2.7
2	0	7.4	0	0.6	6.0	2	3	1.2
3	0	8.2	0	1.5	2.7	2	4	0.05
5	0	1.5	0	3	0.7	4	1.5	1.4
8	0	1.0	0	4	0.2	4	2.5	0.5
10	0	0.5	6	1	1.0
...	6	2	0.05
...	8	2	0.3

to an aluminium-silicon alloy, causes at first an increase in the severity of cracking. With more than 0.5% silicon or 1% copper, the effect is reversed and cracking is decreased. Here also, silicon is more effective than copper, a fact which is apparent from the intervals between the cracking curves in Fig. 1.

From a study of the macrostructures of some of the rings, it was found that those alloys with least silicon and copper were composed entirely of columnar crystals. With increasing alloy content, there was a gradual change towards a finer, equi-axed structure.

(b) *Restrained Welds.*

There is a considerable measure of general agreement between the casting and welding results, although there are one or two differences in detail.

So far as the alloys containing small amounts of copper and silicon are concerned, the onset of cracking is more gradual in Fig. 2 than in Fig. 1. Thus, in the aluminium-silicon binary alloys, the maximum cracking occurs at 0.7% silicon in the restrained-weld tests and at 0.4% silicon in the ring-casting tests. In the aluminium-copper binary

alloys, there is a continuous increase in cracking in the weld test up to

about 3% copper, followed by a steep and then a more gradual decrease.

As in the casting experiments, the addition of copper to aluminium-silicon alloys causes an initial increase in cracking, followed by a gradual reduction. This was found to be true also of additions of silicon to aluminium-copper alloys containing up to about 1% copper, but as no alloys containing 0.5-1% silicon and 1-4% copper were tested, a complete comparison with the casting results is not possible.

Of more practical importance than the results for the most-hot-short alloys are those for the alloys which lie on the outskirts of the hot-short zone. The 1-in. line of cracking, for instance, lies in roughly the same position in each diagram, which suggests that a similar choice of alloys may be made whether for casting or welding, if it is intended primarily to eliminate the tendency to hot-shortness.

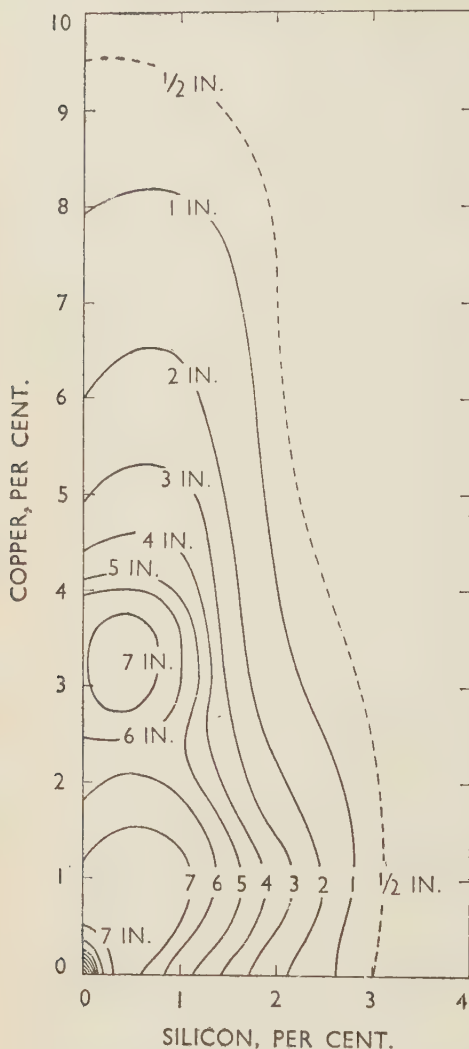


FIG. 1.—Ternary Cracking Diagram for Ring Castings in Aluminium-Copper-Silicon Alloys.

Etching the surfaces of some of the welds showed that, as in the ring castings, the addition of copper and silicon caused a change from a

columnar to a fine-grain equi-axed structure. It also revealed that the cracks and fissures were located at the grain boundaries.

2. Aluminium-Magnesium-Silicon Alloys.

The experimental results for ring castings are given individually in Table III, and graphically in Fig. 3; Fig. 4 is an enlargement of the extreme aluminium corner of Fig. 3. The results of restrained - weld tests, which were carried out on alloys situated on the aluminium - magnesium silicide quasi-binary line, are given in Fig. 5.

(a) Ring Castings.

Figs. 3 and 4 show that in the binary aluminium-magnesium alloys the variation of hot-shortness with composition is similar to that in the systems aluminium-copper and aluminium-silicon, namely an initial increase followed by a more gradual decrease. In no aluminium-magnesium alloy, however, does the severity of cracking approach that of the more hot-short alloys in the other two binary systems.

A similar rise and fall is characteristic of all radial sections through Fig. 3, which suggests that the phenomenon of hot-shortness is basically

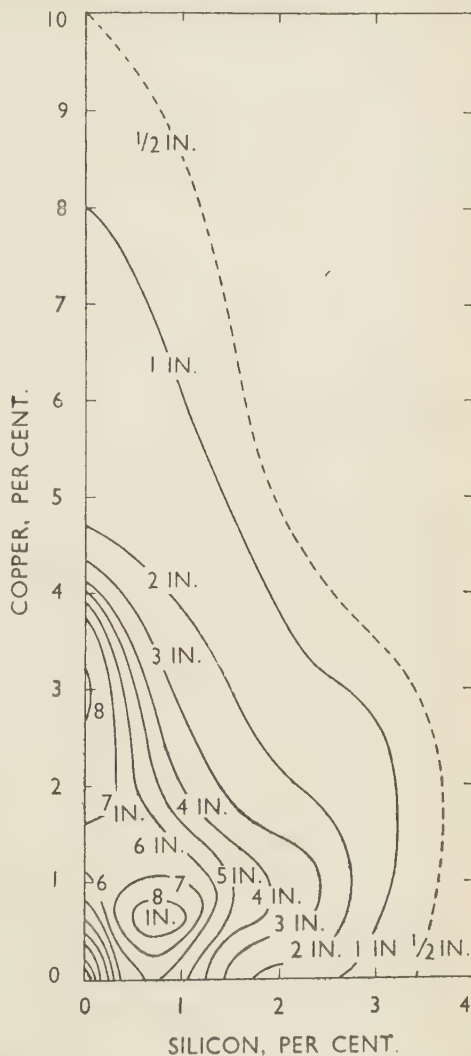


FIG. 2.—Ternary Cracking Diagram for Restrained Welds in Aluminium-Copper-Silicon Alloys.

TABLE III.

(a). Aluminium-Magnesium Alloys.			(b). Mg:Si = 6:1.		
Nominal Composition.		Average Length of Cracking, in.	Nominal Composition.		Average Length of Cracking, in.
Mg, %.	Si, %.		Mg, %.	Si, %.	
0	0	0	0.24	0.04	3.7
0.5	0	2.3	0.49	0.08	4.7
1	0	2.5	0.98	0.16	4.1
2	0	2.5	1.48	0.25	3.5
4	0	1.7	1.97	0.33	3.1
6	0	0.9	2.95	0.49	2.6
8	0	0.5	3.94	0.66	2.1
10	0	0	5.91	0.99	1.4

(c). Mg:Si = 3:1.			(d). Aluminium-Magnesium Silicide Alloys.		
Nominal Composition.		Average Length of Cracking, in.	Nominal Composition.		Average Length of Cracking, in.
Mg, %.	Si, %.		Mg, %.	Si, %.	
0.24	0.08	5.5	0.22	0.13	6.2
0.47	0.16	5.7	0.43	0.25	6.9
0.95	0.32	5.6	0.87	0.50	8.2
1.42	0.47	4.0	1.30	0.75	6.2
1.89	0.63	3.6	1.73	1.00	6.1
2.84	0.95	3.2	2.60	1.50	4.0
3.79	1.26	2.6	3.47	2.00	2.7
5.70	1.90	1.1	5.20	3.00	0.8

(e). Mg:Si = 1:1.			(f). Mg:Si = 1:2.		
Nominal Composition.		Average Length of Cracking, in.	Nominal Composition.		Average Length of Cracking, in.
Mg, %.	Si, %.		Mg, %.	Si, %.	
0.18	0.18	9.2	0.06	0.11	5.2
0.35	0.35	7.4	0.11	0.22	8.0
0.70	0.70	7.9	0.23	0.45	9.1
1.06	1.06	5.9	0.45	0.90	5.5
1.41	1.41	4.6	0.67	1.34	4.2
2.12	2.12	4.1	0.90	1.79	2.5
2.83	2.83	1.2	1.34	2.69	1.1
4.24	4.24	0.3	1.79	3.58	0.3

TABLE III (continued).

(g). Mg:Si = 1:4.		
Nominal Composition.		Average Length of Cracking, in.
Mg, %.	Si, %.	
0.12	0.48	9.0
0.37	1.46	4.5
0.73	2.92	1.1
0.97	3.88	0.5

the same both in binary and ternary systems. There appears to be a maximum hot-shortness at a total alloy content of between 0.5 and 1.0%, and the composition at which cracking becomes negligible shows a gradual variation from about 10%, with a preponderance of magnesium, to about 3% as the silicon axis is approached.

It is also apparent from Fig. 3 that the substitution of magnesium for silicon, or *vice versa*, causes an increase in hot-shortness. The maximum in this direction occurs when the alloying elements are present in the proportions corresponding approximately to the compound Mg_2Si , a feature which is clearly illustrated by the ridge in Fig. 3.

(b) Restrained Welds.

Restrained-weld tests were carried out on six alloys in the quasi-binary aluminium-magnesium silicide system, containing 0.5, 1.0, 1.5, 2, 3, and 6% Mg_2Si . The results are in general agreement with those obtained in the ring-casting test. There is a maximum tendency to cracking at about 1% Mg_2Si , followed by a decrease which is particularly rapid between 1.5 and 2% Mg_2Si .

V.—DISCUSSION.

1. Comparison of the Two Ternary Systems.

The different effects of copper and magnesium in the two ternary systems are clearly illustrated in Figs. 1, 2, and 3.

In Figs. 1 and 2, which relate to aluminium-copper-silicon alloys, the general trend of the curves is from one binary axis to the other, and apart from a bend at each end, their course is moderately straight. This means that, in general, an increase in the copper or silicon content of a ternary aluminium-copper-silicon alloy will reduce the hot-shortness.

In other words, the function of each element is basically the same with regard to the hot-shortness characteristics. Thus, part of the copper

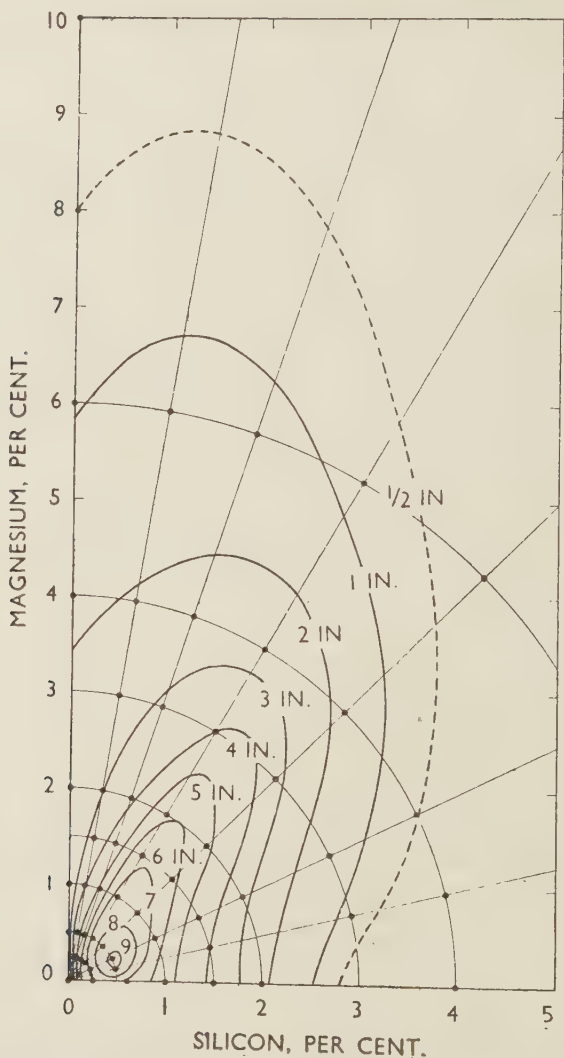


FIG. 3.—Ternary Cracking Diagram for Ring Castings in Aluminium-Magnesium-Silicon Alloys.

may be replaced by silicon without influencing the hot-shortness, provided that it is borne in mind that 1% silicon is roughly equivalent

to 3% copper in its effect on hot-shortness. This ratio is reflected in Figs. 1 and 2 as the gradient of the curves, which is, as a very rough generalization, 3 : 1.

The aluminium-magnesium-silicon alloys are entirely different.

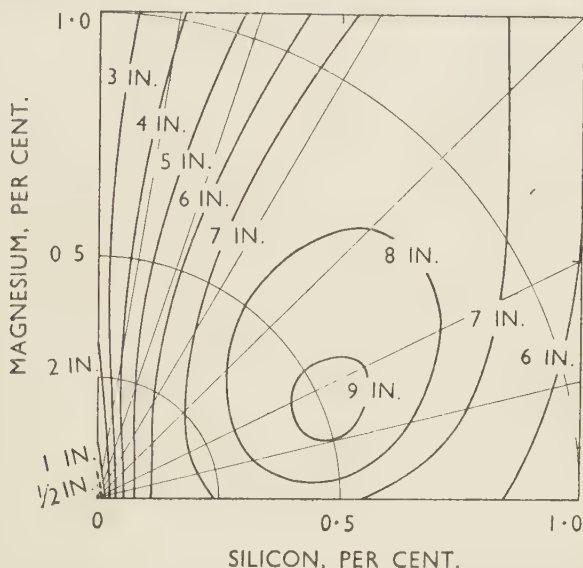


FIG. 4.—Corner of Ternary Cracking Diagram for Ring Castings in Aluminium-Magnesium-Silicon Alloys.

The cracking diagram, Fig. 3, consists essentially of a ridge of relatively severe cracking which occurs at a magnesium : silicon ratio of approximately 2 : 1; and the cracking curves are in general parallel to this

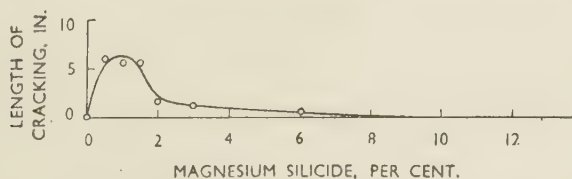


FIG. 5.—The Cracking of Restrained Welds in Aluminium-Magnesium Silicide Alloys.

ridge and have a slope of 2 : 1. This implies that an increase in magnesium content will have the same effect on hot-shortness as a decrease in silicon content, and whether the effect of alloying additions

will be to augment or reduce cracking will depend on which side of the ridge the alloy considered lies. This is the important difference between the two systems. In contrast to copper and silicon together, magnesium and silicon have mutually opposite functions, and consequently an addition of one element to an alloy must be accompanied by an equivalent addition of the other if the resulting alloy is to have similar hot-shortness properties to the original one. The diagram (Fig. 3) shows that 1% silicon is equivalent to about 2% magnesium or, more precisely, to minus 2% magnesium.

Thus, the conditions for a constant degree of hot-shortness in the two systems can be expressed mathematically as follows :

$$(a) \% \text{ Cu} + 3 \times \% \text{ Si} = \text{constant, and}$$

$$(b) \% \text{ Mg} - 2 \times \% \text{ Si} = \text{constant.}$$

These expressions hold only when the slopes of the cracking curves in Figs. 1 and 2 and in Figs. 3 and 4 are 3 : 1 and 2 : 1 respectively, and it will thus be evident that they are rough approximations, based on the trend and not on the exact position of the cracking curves. They are not intended to serve as a basis for predicting the effect on hot-shortness of changes in composition, but merely to emphasize the difference between the two ternary systems. Clearly, the divergence from the ratios of 3 : 1 and 2 : 1 is very considerable in certain parts of the diagrams—particularly for alloys containing less than 1% copper or less than 1% silicon in the aluminium-copper-silicon system, and for alloys in the range 3–10% magnesium, 1–5% silicon in the aluminium-magnesium-silicon system. These ranges, however, are of less industrial importance than those for which the expressions are more nearly true.

2. Comparison with Previous Investigations.

The results are in agreement with the conclusion of Dix and Lyon,² that alloys containing 3–5% copper with 3–5% silicon possess slightly better casting properties than an 8%-copper alloy. Both in the casting and welding tests, the 8%-copper alloy has 1 in. of cracking, while even the most hot-short of the above-mentioned range of ternary alloys gives less than 1 in. The alloy developed by Tichy,³ containing 3% copper and 5% silicon, lies in this range, and his findings regarding its suitability for sand- and die-castings conform with the results of the ring-casting test.

The generally observed tendency towards weld cracking in heat-treatable aluminium-magnesium-silicon alloys would be expected from the present results, since the majority of commercial alloys lie in the

region of intense hot-shortness (Fig. 3), with respect to magnesium and silicon contents. While Fig. 3 relates to casting properties, the few welding tests done on this system indicate a similarity of behaviour during casting and welding.

In the binary alloys, the general reduction in the cracking tendency with increasing copper or silicon content, found by Bochvar and Makimdzhanova⁷ and also by Lees,⁸ is confirmed by the present work. The increase in hot-shortness caused by the first additions of the alloying element was not detected by the above-mentioned investigators, owing to the limited number of alloys they tested. This feature of the hot-shortness properties of binary alloys was noticed by Verö,⁶ whose results are of special interest in relation to aluminium-silicon alloys and were discussed in an earlier paper.¹¹

3. Theoretical Consideration.

(a) Binary Systems.

In a previous paper,¹¹ a theoretical explanation was advanced to account for the hot-shortness phenomena encountered in the binary aluminium-silicon alloys. According to this theory, the condition necessary for the occurrence of hot cracking is the presence in an alloy during freezing of a small proportion of liquid over a sufficiently wide temperature range. In practice, this condition is fulfilled in alloys with a wide freezing range in which the proportion of liquid solidifying at a constant temperature is less than a certain critical value, which for aluminium-silicon alloy castings and welds was found to be about 20%. For binary eutectiferous systems (such as the systems aluminium-silicon, aluminium-copper, and aluminium-magnesium), the first additions of alloying element to pure aluminium cause a large increase in the freezing range and a consequent increase in hot-shortness. Subsequent additions beyond the solid-solubility limit cause no further extension of the freezing range, but increase the proportion of liquid solidifying at the eutectic temperature, thus reducing hot-shortness. The characteristic rise and fall in the curves of hot-shortness against composition can be explained in this way.

The application of the theory to aluminium-silicon alloys was supported by an investigation of the high-temperature tensile properties of these alloys,¹⁵ and it seems likely that the theoretical explanation applies equally well to the aluminium-copper and aluminium-magnesium alloys. This belief is strengthened by the fact that in each of the three systems, cracking becomes negligible in alloys containing the same proportions by volume of eutectic, i.e. of liquid solidifying at a

constant temperature. Calculations of these values for alloys showing 0.5-in. cracking in the ring-casting and restrained-weld tests have been made, assuming the effective solid-solubility limit to coincide with the composition of maximum hot-shortness, and the relevant data are set out in Table IV. The conversion from weight percentage to volume percentage has been calculated according to the mixture law from the densities of the elements.

TABLE IV.

System.	Effective Solid-Solubility Limit, wt.-%.	Equilibrium Eutectic Composition, wt.-%.	Alloy Giving 0.5-in. Cracking.		
			Composition, wt.-%.	Eutectic, %.	
				By Weight.	By Volume.
Al-Si	0.4	11.7	3	23.0	23.4
Al-Cu	1.0	33	10	28.1	23.2
Al-Mg	1.5	35	8	19.4	22.3

The proportion by volume of liquid freezing at a constant temperature which is necessary to prevent cracking would be expected to vary according to the conditions of the test and the structure of the casting or weld, though for a given test and in constant conditions the equality between the values for different alloy systems should be maintained. In this connection, it is interesting to observe that in Verö's casting test the critical proportion of liquid was 12-13%.

There are one or two points, however, which the theory does not explain, such as the existence of the two maxima in the hot-shortness curve for aluminium-copper alloys, and the relatively slight cracking of aluminium-magnesium alloys in spite of their extended freezing range.

(b) *Ternary Systems.*

It is reasonable to assume that the hot-shortness of a ternary alloy, in practical conditions of casting and welding, depends upon the existence and extent of the temperature range above the solidus in which it possesses little strength and no elongation, as was found to be the case with binary aluminium-silicon alloys. It should therefore be possible to predict the hot-shortness of a ternary alloy from a knowledge of the properties of the binary alloy nearest in composition and of the effect on the constitution of additions of the third element.

(c) *Aluminium-Copper-Silicon Alloys.*

First consideration may be given to the aluminium-silicon alloys containing a proportion of eutectic, in which freezing is complete at

577° C. According to Gwyer, Phillips, and Mann,¹³ the first additions of copper cause a rapid decrease in the temperature at which freezing is complete from 577° to 525° C., the ternary eutectic temperature. This decrease occurs over a composition range the extent of which depends on the solid solubility of copper in the aluminium phase and in the binary aluminium-silicon complex. This solubility is stated by the above-mentioned workers as not exceeding 0.25%. As a result of this decrease in the solidus temperature, an increase in hot-shortness is to be expected, and is in fact observed (Figs. 1 and 2). Similarly, the first additions of silicon to aluminium-copper alloys cause a decrease in the solidus temperature from 548° to 525° C. The decrease here, however, is only 23° C., as compared with 52° C. in the previous case, and the increase in hot-shortness would therefore be expected to be less marked, as is found in the ring-casting experiments. (In the welding experiments, no increase was found in the hot-shortness of aluminium-copper alloys with additions of silicon, for the reason already advanced.)

After the initial decrease in solidus temperature, further additions of the third element can have no effect in this direction. Their effect is to reduce the liquidus temperature, thereby shortening the freezing range, and to increase the proportion of ternary eutectic. A reduction in the hot-short temperature range may be expected, with a corresponding decrease in hot-shortness. An explanation of the effect of larger additions of copper or silicon in reducing hot-shortness is thus possible.

(d) *Aluminium-Magnesium-Silicon Alloys.*

The explanation of the hot-shortness properties of the aluminium-magnesium-silicon alloys is somewhat involved. This is to be expected since these alloys are constitutionally more complex than the aluminium-copper-silicon alloys because of the compound Mg_2Si , which gives rise to two ternary eutectics whose freezing points differ by 100° C. It is, however, of interest to apply to these alloys the theoretical treatment applied to the aluminium-copper-silicon alloys. The constitutional data for this treatment are provided by the work of Phillips¹⁴ except where otherwise stated, and apply to alloys in the metastable condition.

The addition of a small quantity of magnesium to an aluminium-silicon alloy lowers the solidus temperature from 577° to 551° C. An increase in hot-shortness would then be expected, and is in fact found, there being an increase in cracking from 7 to 9 in. when 0.2% magnesium is added to a 0.5%-silicon alloy. At higher percentages of silicon, the increase in cracking is still definite, although more gradual.

A prediction of the effect on hot-shortness of additions of silicon to aluminium-magnesium alloys is complicated by the uncertainty of the

extent of the freezing range in the binary system, which depends on the effective value of the solid-solubility limit. Phillips¹⁴ places the limit of solid solubility at 4.25% magnesium in metastable conditions, but the position of the hot-shortness maximum in the present work suggests that Al_3Mg_2 makes its first appearance between 1 and 2% magnesium at the rates of cooling that obtain in the ring-casting test. It would, therefore, be unwise to make any prediction of the effect of additions of silicon to alloys containing less than 4.25% magnesium. At magnesium contents above 4.25%, there is only a small degree of hot-shortness, and no great importance can be attached to the slight increase which is effected by silicon additions. From considerations of the extent of the freezing range, however, a decrease in hot-shortness would be expected to accompany an increase in silicon content because there is a gradual increase in the solidus temperature from the 450° C. ternary-eutectic plane to the 551° C. plane. The reason for this discrepancy is not apparent, but factors other than freezing range may possibly influence the extent of the hot-short temperature range, such as, for instance, the temperature at which secondary separation begins.

There now remains to be considered the pronounced tendency to hot-shortness associated with the aluminium-magnesium silicide quasi-binary line. An examination of the extent of the freezing range offers no clue to the reason for this hot-shortness ridge, since the liquidus surface is smooth and the solidus is a plane surface at 551° C. The surface of secondary separation does, however, ascend to a maximum at 594° C. on a line which coincides roughly with the quasi-binary section. It seems that this is likely to lead to a condition during freezing in which a small proportion of liquid is present over a relatively large temperature range in alloys lying on or near the quasi-binary line, compared with alloys further away from this line. Any such condition would be expected to lead to a greater degree of hot-shortness the nearer the composition approaches the magnesium silicide ratio.

This tentative explanation finds support in the work of Hanson and Gayler.¹⁶ From cooling-curve data, they concluded that the aluminium-magnesium silicide alloys formed a eutectiferous system, the eutectic temperature being about 590° C. They noticed a further slight arrest on cooling at about 550° C., which they ascribed to the presence of traces of ternary eutectic, arising from a slight discrepancy in composition. Phillips, however, found that the reaction between magnesium and silicon was a balanced one, resulting in the presence of free silicon, even in alloys lying on the quasi-binary line. This free silicon would lower the solidus to the ternary-eutectic plane and doubtless is the reason for the arrest at 550° C. found by Hanson and Gayler.

The existence in the aluminium-magnesium silicon alloys of the quasi-binary series suggests that the two systems Al-Mg-Mg₂Si and Al-Mg₂Si-Si might be considered separately. In the latter system, for instance, it can be seen from Fig. 3 that the cracking curves resemble fairly closely those in the aluminium-copper-silicon system. There is a maximum in the ternary field near the aluminium corner, as well as on each binary axis. After this initial increase, there is a gradual decrease in hot-shortness as the alloy content is raised.

VI.—CONCLUSIONS.

The hot-shortness of aluminium-rich alloys in the aluminium-copper-silicon and aluminium-magnesium-silicon systems has been investigated by a casting test and a restrained-weld test. The first alloying additions to pure aluminium cause a rapid increase in hot-shortness which, after passing through a maximum, gradually decreases as the alloy content is raised.

In the aluminium copper-silicon system, additions of silicon are more effective than additions of copper in reducing hot-shortness once the maximum hot-shortness has been passed. In the aluminium-magnesium silicon system, the most-hot-short alloys are those near the quasi-binary aluminium magnesium silicide line. The presence of silicon or magnesium in excess of the Mg₂Si ratio causes, in general, a reduction in cracking.

The hot-shortness of binary and ternary alloys appears to depend upon the extent of the hot-short temperature range, i.e. that range in which the alloys possess appreciable strength but no elongation; this in turn is a function of the freezing range and the proportion of liquid freezing at a constant temperature. In ternary systems, however, other factors, such as the temperature at which secondary separation begins, may be of importance. The application of the theory of hot-shortness to aluminium-magnesium-silicon alloys is complicated by the presence of the magnesium silicide phase.

Reasonably good mechanical properties are possessed by those aluminium-copper silicon alloys which are likely to be of use from the point of view of hot-shortness.

ACKNOWLEDGEMENTS.

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APPENDIX I.

THE TENSILE PROPERTIES OF SOME ALUMINIUM-COPPER-SILICON ALLOYS.

In order to supplement the work on the hot-shortness of aluminium-copper-silicon alloys, six alloys having the most promising hot-shortness properties were selected for mechanical testing.

Tensile and hardness tests were carried out on each of the six alloys in the form of 14-gauge sheet, both in the annealed and heat-treated conditions. The tensile properties in the welded condition were determined by testing the original restrained welds made in annealed material, and also a further set of welds made in heat-treated material.

A standard heat-treatment was applied to all six alloys, consisting of a solution treatment for 5 hr. at 515° C., followed by water quenching and subsequent ageing for 12 hr. at 160°–165° C. Tensile test-pieces were machined to conform with the requirements of British Standard Specification No. 485. In the case of welded material, the weld was in the centre of the parallel portion of the test-piece and transverse to the length, and the excess weld metal was removed. The results, which are given in Tables V and VI, are the average of either three or four tests in every case.

During tensile testing, the welded specimens made from annealed material generally fractured in the basis metal, indicating a higher strength in the cast condition than in the annealed. As would be expected, when such a fracture occurred the strength obtained approximated to that of the annealed material; but the values for elongation

were much lower, owing presumably to the higher proof stress of the weld metal confining the elongation to the remainder of the test-piece.

In the heat-treated condition, all the alloys had good welding pro-

TABLE V.—*Mechanical Properties of Aluminium-Copper-Silicon Alloys.*

Composition.		Results of Tests on Annealed Material.				Results of Tests on Heat-Treated Material.			
Cu, %.	Si, %.	U.T.S., tons/in. ² .	0.1% Proof Stress, tons/in. ² .	Elongation, %.	Hardness, D.P.H.	U.T.S., tons/in. ² .	0.1% Proof Stress, tons/in. ² .	Elongation, %.	Hardness, D.P.H.
0	4	6.7	2.1	24.5	30	9.5	4.5	20	53
8	0	8.85	2.5	18.5	39.5	25.6	14.4	12	122.5
2	3	10.0	2.4	22.5	41	14.5	6.25	16	75
2	4	9.2	2.6	22.5	40	15.0	7.55	16	75
4	1.5	9.2	2.65	26.5	39	24.5	14.5	9.5	124.5
6	2	10.4	2.8	24	43.5	26.1	16.5	13	128

TABLE VI.—*Mechanical Properties of Welds in Aluminium-Copper-Silicon Alloys.*

Composition.		Results of Tests on Welds Made in Annealed Material.				Results of Tests on Welds Made in Heat-Treated Material.		
Cu, %.	Si, %.	Average Length of Cracking in Restrained-Weld Tests, in.	U.T.S., tons/in. ² .	Elongation, %.	Position of Fracture.	U.T.S., tons/in. ² .	Elongation, %.	Position of Fracture.
0	4	0.2	6.6	18.5	in basis metal	8.0	14	in basis metal.
8	0	1.0	10.1	10.5	in basis metal	13.25	5	in weld.
2	3	1.2	10.1	9	in weld	10.8	10	in weld.
2	4	0.05	10.9	13	in basis metal	11.4	8	in weld or weld area.
4	1.5	1.4	10.3	11.5	in basis metal	12.25	6.5	in weld.
6	2	0.05	10.1	6	in basis metal	10.9	3	in weld or at edge of weld.

perties, particularly those with high copper content, which are especially susceptible to heat-treatment. Welding brought about a great reduction in strength, owing to the presence of the cast structure in the weld, but in this condition the alloys were appreciably stronger than after welding in the annealed condition. The values for elongation, however,

were lower than for the heat-treated basis metal, and were also lower than those for welds in the annealed material. This was probably due to the high relative strength of the basis metal having the effect of localizing elongation and fracture to the vicinity of the weld.

Where high strength in the heat-treated condition, together with reasonably good welding properties, are required, material based on the 4%-copper, 1.5%-silicon alloy should give the most satisfactory results. Where good welding properties are the first consideration, but moderate strength is also desirable, alloys of the 2%-copper, 4%-silicon type would seem most suitable.

A CONSIDERATION OF THE CONSTITUTION 1120 OF ALUMINIUM-IRON-SILICON ALLOYS AND ITS RELATION TO CRACKING ABOVE THE SOLIDUS.*

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SYNOPSIS.

An earlier paper, "Hot-Shortness of Some Aluminium-Iron-Silicon Alloys of High Purity", left unanswered the question whether the susceptibility to cracking was related to the constitution, and if it was, in what manner.

In order to establish the existence of any such relationship, reference has been made to the literature concerning the constitution of the aluminium-rich alloys in the aluminium-iron-silicon system. From this it has become apparent that in alloys containing more iron than silicon the theoretical solidus temperature is not evidenced in practice, the completion of freezing occurring at a much higher temperature. Upon this fact is based an explanation of the phenomena generally classed under the heading of hot-shortness, similar in essentials to that previously established for binary aluminium-silicon alloys.

I.—INTRODUCTION.

THE investigation of the hot-shortness of alloys of aluminium with silicon and with silicon and iron together has formed the basis of two earlier papers.^{1, 2} In the first of these, a theory was advanced relating the results of the experimental work on the aluminium-silicon alloys to their constitution, and it was suggested that this theory might be generally applicable to the cracking of aluminium alloys at temperatures above the solidus. At the time of publication of the results for aluminium-iron-silicon alloys, the correlation with the constitutional diagram was, for a variety of reasons, far from complete; but after a more exhaustive study of the literature relating to equilibrium and metastable conditions in this system, the position has become considerably clearer. It has thus been possible to show that the theory of hot-short cracking is, with a few modifications, of wider application than was at first indicated by the work on aluminium-silicon alloys.

While the constitution of aluminium-silicon alloys is simple and is

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TABLE I.—*The Constitution of Aluminium-Rich*

Date.	Investigators.	Ternary Phases in Aluminium-Rich Aluminium-Iron-Silicon							
		α -(Fe-Si) (Dix and Heath).							
		Designation.	Composition.		Formula.	Mode of Formation.	Invariant Point.		
Fe, %.	Si, %.		Fe, %.	Si, %.			° C.		
1921	Rosenhain, Archbutt, and Hanson ³								
1921	Merica, Waltenberg, and Freeman ⁴								
1927	Gwyer and Phillips ⁵	β	42.5 — 25 (Variable)	18		$\text{FeAl}_2 + \text{liq.} \rightarrow \beta$	2.0	4.0	629
1928	Dix and Heath ⁶	α - (Fe-Si)	30	8					
1931	Fink and Van Horn ⁷	Solid solution of silicon in FeAl_3							
1931	Fuss ⁸								
1933	Nishimura ⁹								
1936	Jäniche ¹⁰								
1937	Urazov and Shashin ¹¹								
1937	Sergeev and Rimmer ¹²								
1940	Takeda and Mutuzaki ¹³	K_5	38.1	15.9	$\text{Al}_{15}\text{Fe}_6\text{Si}_5$	$\text{FeAl}_2 + \text{liq.} \rightarrow \text{K}_5$	2	3	620
1943	Phillips and Varley ¹⁴	α - (Fe-Si)	Variable			$\text{FeAl}_3 + \text{liq.} \rightarrow$ α -(Fe-Si)	2.0	4.0	629

known with accuracy, that of aluminium-iron-silicon alloys is of much greater complexity; and although it has been the subject of numerous investigations, several points remain to be clarified. For this reason, it was thought unsatisfactory to select any single work on which to base the discussion of the hot-shortness, and the conclusions of the more comprehensive researches have therefore been compared.

II.—THE CONSTITUTION OF ALUMINIUM-RICH ALUMINIUM-IRON-SILICON ALLOYS.

In practical conditions of casting, welding, &c., commercial-purity aluminium (i.e. aluminium containing variable quantities of iron and

Aluminium-Iron-Silicon Alloys.

Alloys (Containing Less than 13% Silicon).							Ternary Eutectic.				
β -(Fe-Si) (Dix and Heath).											
Designation.	Composition.		Formula.	Mode of Formation.	Invariant Point.			Constituent Phases.	Invariant Point.		
	Fe, %.	Si, %.			Fe, %.	Si, %.	° C.		Fe, %.	Si, %.	° C.
X				$\text{FeAl}_3 + \text{liq} \rightarrow \text{X}$			ca. 605	$\text{Al} + \text{X} + \text{Si}$			
X							610				
X	Variable			$\beta + \text{liq.} \rightarrow \text{X}$	1.5	7.5	611	$\text{Al} + \text{X} + \text{Si}$	0.8	11.6	577
β -(Fe-Si)	27	15									
β -(Fe-Si)	Variable										
	31.2	23.5	$\text{Al}_3\text{Fe}_2\text{Si}_3$					$\text{Al} + \text{Al}_3\text{Fe}_2\text{Si}_3 + \text{Si}$	0.5	12.5	577
T	46	25	$\text{Al}_3\text{Fe}_2\text{Si}_3$	$\text{FeAl}_3 + \text{liq.} \rightarrow \text{T}$	1	7	615	None			
	25.4	25.5	Al_4FeSi_2	$\text{FeAl}_3 + \text{liq.} \rightarrow \text{Al}_4\text{FeSi}_2$	2.7	5.2	645	$\text{Al} + \text{Al}_4\text{FeSi}_2 + \text{Si}$	0.4	11.3	573
β	Variable				1.4	5.0	635	$\text{Al} + \beta + \text{Si}$	0.5	12.0	566
x	29.1 -25.5	14.6 -12.8	Al_4FeSi or Al_3FeSi		2.2	5.8	...	$\text{Al} + \text{x} + \text{Si}$	0.7	12.0	...
K_s	29.1	14.6	Al_4FeSi	$\text{K}_s + \text{liq.} \rightarrow \text{K}_s$	2	5	615	None			
β -(Fe-Si)	Variable			$\alpha\text{-(Fe-Si)} + \text{liq.} \rightarrow \beta\text{-(Fe-Si)}$	1.5	7.5	611	$\text{Al} + \beta\text{-(Fe-Si)} + \text{Si}$	0.8	11.6	577

silicon) is far from being in structural equilibrium. Experimental work on the constitution must, therefore, have been carried out in similar conditions of metastability if the constitutional results are to be wholly applicable to the explanation of the hot-shortness. Most of the constitutional data which have been published do, in fact, refer to conditions of metastability, having for the greater part been obtained by cooling-curve determinations and by microscopic examination of cast structures quenched from high temperatures. In some cases, however, the precise conditions of experiment have not been specified, and the value of the results has thus been lessened.

Some of the results of the more far-reaching investigations are summarized in Table I, which includes the results of several investi-

gations relating to the number and nature of the ternary phases present in alloys lying on the aluminium side of the ternary eutectic. In this Table, all compositions are expressed as weight per cent.

It is not the purpose of the present paper to attempt a compilation of the large amount of available information concerning the constitution of aluminium-iron-silicon alloys, or to arrive at a definite conclusion regarding the several points which are in dispute, although this would undoubtedly be of value; but the following summary of the salient features of the constitutional diagram in metastable conditions is included for the purpose of interpreting the hot-shortness results.

1. *The Liquidus.*

The form of the liquidus surface, as originally determined by Gwyer and Phillips,⁵ has been revised by Phillips and Varley,¹⁴ whose liquidus diagram for alloys containing up to 12% silicon and 6% iron is apparently the only one of its kind from which accurate information can be gained. In the primary aluminium-phase field, the addition of silicon causes a gradual fall in the liquidus temperature, but iron, on the other hand, has little effect, the isothermals lying roughly parallel with the aluminium-iron axis. In the phase fields in which the primary separation is that of FeAl_3 or a ternary phase, an increase in the iron content causes a steep rise in the liquidus temperature.

2. *The Secondary Surface.*

Although few investigators have published diagrams of the surface of secondary separation, most have indicated the positions of the binary valleys. Since in alloys in which aluminium is the primary constituent, the temperature at which secondary separation begins depends principally upon the position of the valley lines with respect to composition and temperature, a comparison between the various researches is possible. Such a comparison is afforded by the data given for the invariant points in Table I, which indicate that, with a few exceptions, a reasonable degree of correspondence exists between the figures obtained by the various investigators.

3. *The Solidus.*

A theoretical consideration of the process of solidification leads to the conclusion that all ternary alloys of compositions lying beyond the limits of solid solubility contain a proportion of liquid down to the ternary-eutectic temperature. On most diagrams the solidus is, therefore, indicated as occurring at about 577°C. , as originally postulated by Gwyer and Phillips,⁵ although Varley, in private correspon-

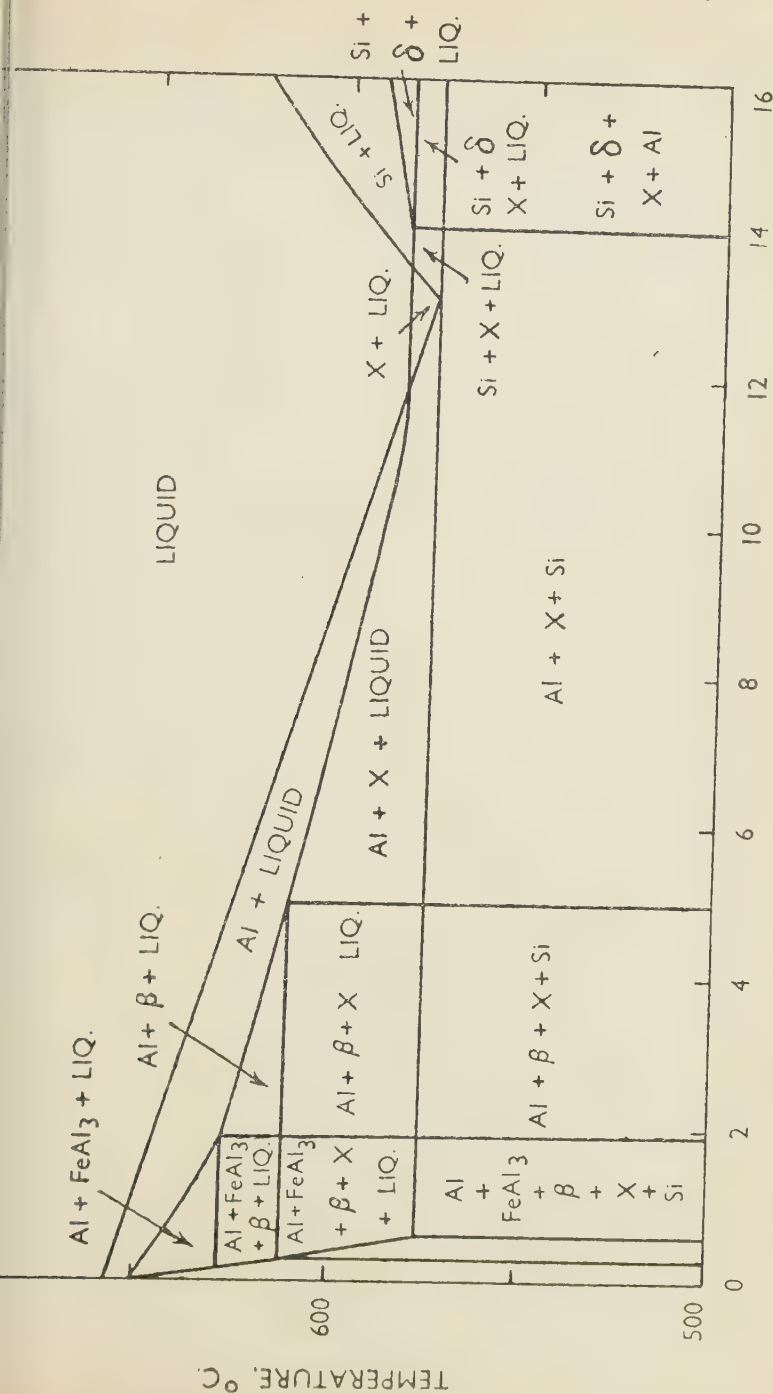


Fig. 1.—Arrests and Constituents. Alloys containing 1.0% Iron. (Gwyer and Phillips.)

dence, states that with increasing quantities of iron the eutectic arrest is of progressively shorter duration, eventually disappearing altogether. Nishimura⁹ and other investigators,¹³ failing to detect the presence of a ternary eutectic, have given diagrams which appear to be more in conformity with the practical observations regarding the position of the solidus, and in which there is a more or less gradual rise in the solidus temperature from 577° C. to about 615° C. as the iron : silicon ratio approaches and exceeds 1 : 1. These two conceptions of the mode of solidification are illustrated by Figs. 1 and 2, taken respectively from the work of Gwyer and Phillips and of Nishimura.

4. *Solid Phases in Aluminium-Rich Alloys.*

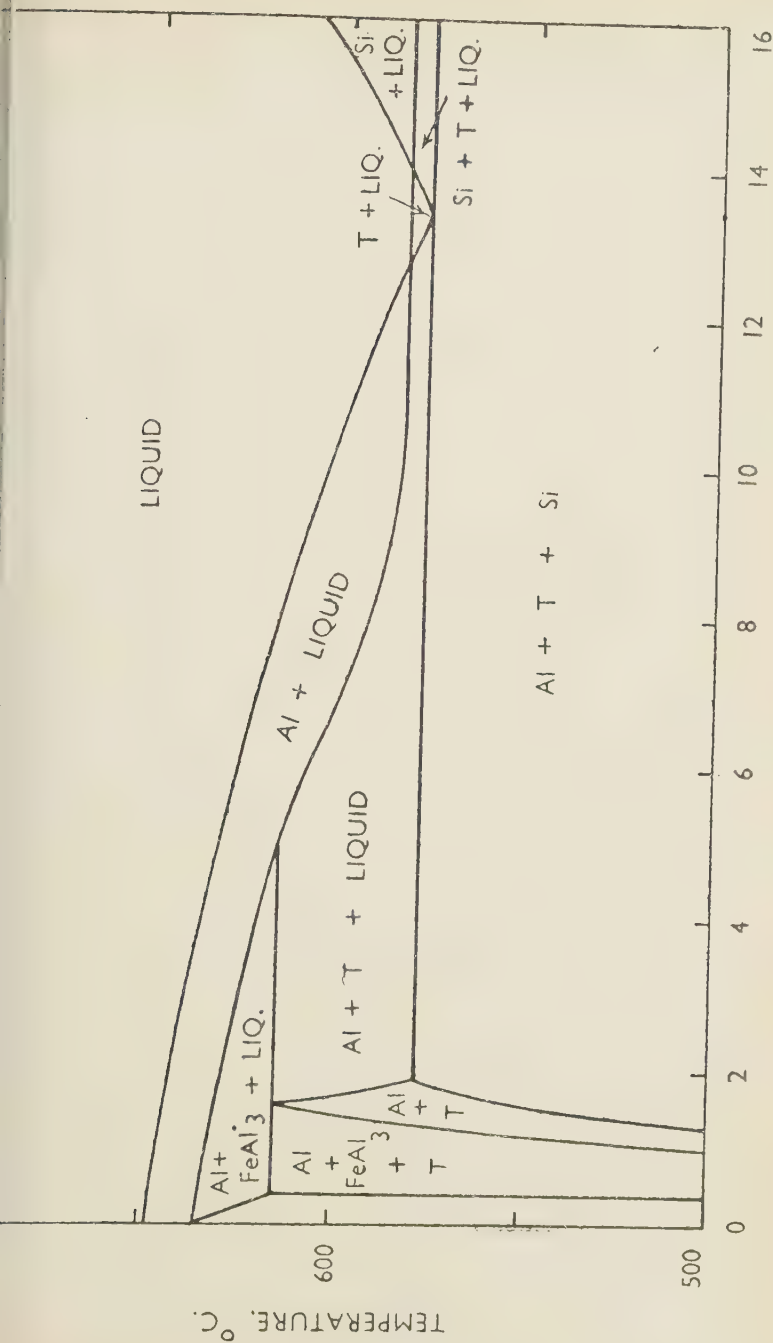
With the exception of Takeda and Mutuzaki,¹³ who mention a compound FeAl_2 , it is generally agreed that binary aluminium-iron alloys contain the phase FeAl_3 . While it has been stated that FeAl_3 is capable of dissolving considerable quantities of silicon, the bulk of the evidence supports the view that the smallest additions of silicon to a binary aluminium-iron alloy cause the formation of a ternary phase, α -(Fe-Si). It is agreed that larger additions of silicon cause the appearance, first of the phase β -(Fe-Si) and finally of free silicon, and a diminution in the amounts of FeAl_3 and α -(Fe-Si). In metastable conditions, all five phases—aluminium, FeAl_3 , α -(Fe-Si), β -(Fe-Si), and silicon—can exist together in certain alloys.

Diagrams showing the constituents present in as-cast alloys have been prepared by Gwyer and Phillips,⁵ whose results are given in Fig. 3, and by Sergeev and Rimmer¹²; and as far as a comparison is possible, these diagrams are in substantial agreement.

III.—THE RELATION BETWEEN CONSTITUTION AND CRACKING ABOVE THE SOLIDUS.

The results obtained from the ring-casting and restrained-weld tests with aluminium-iron-silicon alloys can be seen by reference to the original paper.² It was found that the effect on the cracking, of additions of iron to aluminium-silicon alloys, is to inhibit the influence of the silicon. Thus, in the presence of iron, the composition at which silicon causes the onset of cracking is raised from the extremely low value obtained in the absence of iron to a percentage roughly equal to that of the iron present. The addition of iron causes the maximum cracking to occur at higher percentages of silicon, and when more than 0.1-0.2% iron is present, it has the effect of decreasing the maximum amount of cracking.

In an earlier paper, it was found possible to associate the tendency



SILICON, PER CENT.

FIG. 2.—Arrests and Constituents. Alloys containing 1.0% Iron. (Nishimura.)

to cracking of aluminium-silicon alloys with the extent of the "hot-short temperature range", which was in turn dependent principally upon the freezing range and the proportion of liquid freezing at constant temperature. Thus the extent of the cracking could be related to these two basic factors. A similar correlation would appear to obtain in the aluminium-iron-silicon alloys.

The results for alloys containing 0.2% iron are typical of the whole, and a general explanation will therefore be attempted by a particular consideration of the effect of successive additions of silicon to an aluminium-iron alloy containing 0.2% iron.

This alloy has a freezing range of only a few degrees, but the addition of silicon causes a fall in the solidus temperature, first to 615° C. and, with larger additions, to 577° C. As has already been pointed out, it appears that the fall from 615° to 577° C. occurs at an iron : silicon ratio of about 1 : 1, i.e. when the silicon content is of the order of 0.2%. With increase in the silicon content of the 0.2%-iron alloy, the curve showing the tendency to cracking would thus be expected to rise, slowly at first, from the very low value for the binary alloy, and then more rapidly with the occurrence of the second fall in the solidus temperature. Inspection of the appropriate curve in Fig. 9 of the original paper² indicates that this expectation was, in fact, realized. There is a gradual increase in the severity of cracking up to 0.25% silicon, followed by a much more rapid increase between 0.25 and 0.4% silicon. The reduction caused by larger silicon additions is attributable to their effect in increasing the proportion of liquid freezing at the eutectic temperature.

The relation between constitution and cracking above the solidus may be viewed from a different angle by comparing the freezing range for the alloys in the extreme aluminium corner of the ternary constitutional diagram with the ternary diagram of cracking. The fall in the solidus temperature from 615° to 577° C. occurs at an iron : silicon ratio of approximately 1 : 1, and the corresponding increase in the freezing range therefore coincides with the onset of appreciable cracking.

The thermal data which indicate that the true solidus is not observed in alloys in which there is a preponderance of iron over silicon, are supported by microscopic evidence. Fig. 3, taken from the paper by Gwyer and Phillips,⁵ shows the constituents present in as-cast alloys at room temperature, and indicates that FeAl_3 makes its appearance when the iron : silicon ratio exceeds 1 : 1. In the discussion of the original paper on the hot-shortness of aluminium-iron-silicon alloys, Varley pointed out that although free silicon was present on either side of the line which lies at an iron : silicon ratio of about 1 : 1, it existed in much smaller quantities in the presence of FeAl_3 than in the

of aluminium-iron-silicon alloys at temperatures above the solidus is in part attributable.

IV.—CONCLUSIONS.

The relationship between the constitution and the susceptibility to cracking of aluminium-rich aluminium-iron-silicon alloys appears to be similar to that suggested for binary aluminium-silicon alloys. This relationship is, however, less easily established in the ternary system because of the uncertainty regarding particular features of the constitutional diagram.

The small tendency to cracking of alloys in which there is a preponderance of iron over silicon is apparently due to the fact that, as the iron content increases, the ternary-eutectic arrest at 577° C. diminishes in magnitude and becomes unobservable at an iron : silicon ratio of about 1 : 1. The major part of the solidification is thus completed at about 615° C., and there is an effective shortening of the liquidus-solidus interval. This contention has been supported by the microscopic evidence of other investigators, which provides the additional information that when the total silicon content is less than the iron content, the free silicon is present in a form which renders it ineffective in promoting a tendency to cracking at temperatures above the solidus.

ACKNOWLEDGEMENTS.

The authors wish to acknowledge their indebtedness to Professor L. Aitchison, D.Met., under whose general direction the work has been conducted, to Major P. C. Varley for information relating to the constitution of the alloys, and to the Aluminium Development Association, who sanctioned the publication of this paper.

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PRESSURE AND CREEP TESTS AT CONSTANT 1121 HOOP STRESS ON LEAD AND ALLOY "E" PIPES.*

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SYNOPSIS.

A wide range of tests in connection with pressure-cable sheath problems is in progress in the laboratories of British Insulated Callender's Cables, Ltd., on lead and lead alloy pipes subjected to internal pressure, the pressure being kept constant for each test. A method of testing the pipes at constant hoop stress, necessitating pressure adjustments, is described; and from the results, the Andrade creep constants β and k , considered to represent two different types of creep flow, have been determined. The results would appear to throw some light on the "extensibility characteristics" disclosed by the constant-pressure tests, which are briefly reviewed. These show a greatly diminished extension before fracture for alloy "E", but not (so far) for pure lead at slow creep rates. Some consideration is given to the nature of creep flow, and a hypothesis is developed to account for the extension results. Applications to some problems connected with the use of lead sheath for high-voltage pressure cables are dealt with.

I.—INTRODUCTION.

A LARGE number of tests has been carried out and is still in progress at the Wood Lane Research Laboratories and at the Technical Control Laboratories of British Insulated Callender's Cables, Ltd., on lead and lead alloy pipes of cable-sheath size, to determine the extension obtained before fracture at different rates of creep under internal pressure. The pressures used have ranged from a few to the order of 150 lb./in.² (the latter for short-time tests with the harder alloys), corresponding to hoop stresses ranging from about 100 to 1500 lb./in.², and each test has been carried out under conditions of constant pressure. This leads to approximately constant creep rate over most of the test period, and is analogous, although not exactly so, to constant-load creep testing, in the ordinary way, on specimens of strip, &c.

These tests are also analogous to constant-load creep tests in that the applied stress increases as creep progresses. The hoop stress f , produced in the wall of a pipe of internal diameter d_i and thickness b by internal pressure P is, using the formula for thin-walled pipe, given by: $f = Pd_i/2b$. As creep progresses, d_i increases and b decreases; consequently, f increases if the pressure is constant.

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The advantages of creep tests under constant-stress conditions have been pointed out by Andrade,^{1, 2, 3} and he has derived an expression for creep rate under such conditions. This expression is particularly useful in that it enables determinations to be made of constants believed to be associated with two distinct types of creep flow. The importance of such determinations will be discussed, and a full description of the Andrade equation is given in Section V.

An expression has been derived whereby the hoop stress can be kept practically constant during a pipe-pressure test by suitable adjustment of the internal pressure; and measurements have been made on nominally pure lead, alloy "E" (antimony 0.2, tin 0.4%, lead remainder, to British Standard Specification No. 801), and on several other alloys. Only the work on pure lead and alloy "E" is described in the present paper; it is hoped to publish the results on other alloys in due course. The application of the results to the particular problem of sheath for pressure cable will be dealt with in detail.

II.—THE PRESSURE/HOOP-STRESS RELATIONSHIP.

The thin-walled-tube relationship given above is usually considered sufficiently accurate if $\frac{b}{d_i} < \frac{1}{20}$. In the present work, two main pipe sizes are concerned: $d_i = 2.76$, $b = 0.12$ in.; and $d_i = 1.8$, $b = 0.1$ in., respectively. The application of the usual relationship is, therefore, less accurate for the second size than for the former. It is probable, however, that the Lamé thick-wall formula is no more accurate in application, as it is based on the assumption of elastic strain, which results in a stress gradient in the wall which probably differs considerably from that which occurs under creep conditions. The simple thin-wall formula has, therefore, been used, and the relationship between hoop stress and diameter under conditions of creep has been derived on the assumption of constant cross-sectional area of the metal. The justification for this assumption is that the change of pipe length is practically negligible; constant cross-sectional area follows then from constant volume.

The cross-sectional area of the metal is given by: $a = \pi d_m b$, where d_m = the mean of internal and external diameters. Substituting in the thin-wall-pipe formula gives:

$$f = P\pi d_i d_m / 2a,$$

and if d_0 = external diameter

$$f = \pi P(d_0 - 2b)(d_0 - b)/2a$$

This simplifies to :

$$f = \frac{P}{2} \left(\frac{\pi d_0^2}{a} - 3 - \frac{b}{d_m} \right)$$

With the present pipe sizes, the error in omitting b/d_m is very slight (less than $\frac{1}{2}\%$), so it is sufficiently accurate to take :

$$f = \frac{P}{2} \left(\frac{\pi d_0^2}{a} - 3 \right)$$

and since $a = \pi d_m' b'$, where $d_m' = \text{original mean diameter}$, and $b' = \text{original wall thickness}$, it is simple to obtain the expression :

$$P = \frac{2fd_m'b'}{q},$$

where

$$q = d_0^2 - 3d_m'b'.$$

From this expression, by periodic measurements of d_0 , the value of P required to keep f constant can readily be obtained. It has been found convenient for calculation purposes to make use of the straight-line relationship between P and $1/q$. The value of q , and hence its reciprocal, is obtained for different values of d_0 , and the corresponding value of P , for given hoop stress, is read off from the straight-line diagram.

III.—MATERIALS USED AND EXPERIMENTAL METHOD.

The materials used in this investigation were high-purity lead pipe and alloy "E" pipe, produced on a continuous-extrusion machine used for cable sheathing at the Prescott Works of British Insulated Callender's Cables, Ltd. The material was produced as pipe (not sheath) in the two sizes given above, for experiments on fracture by creep under internal pressure, with particular reference to the use of these materials as sheath for impregnated gas pressure cable. As mentioned above, these experiments formed part of a large number on pure-lead and alloy pipes produced on continuous and ram-type cable-sheathing presses. A brief review of some outstanding results on pure lead and alloy "E", so far as they concern the present work, is given below.

An analysis of one sample of the pure lead gave this result : copper 0.0005, zinc 0.00104, antimony 0.0021, tin 0.00045, cadmium 0.0001, nickel 0.0001, bismuth 0.0001, iron 0.00035, and arsenic 0.0006%. Analysis of the alloy "E" gave : tin 0.386, and antimony 0.253%. Minor constituents were not determined in this alloy.

Results of grain-size determinations are given in Table I. These are averages of a large number of measurements on ring samples, and were determined by dividing the number of grains, counted radially, by the wall thickness at four different localities (12, 3, 6, and 9 o'clock positions) on each ring.

TABLE I.

Material.	Pipe.	Grain-Size, mm.		
		Mean.	Maximum Individual.	Minimum Individual.
Pure lead.	3 in. outside dia. \times 0.12 in. wall.	0.74	1.6	0.43
Pure lead.	2 in. outside dia. \times 0.1 in. wall.	0.58	1.3	0.42
Alloy "E".	3 in. outside dia. \times 0.12 in. wall.	0.33	0.51	0.22
Alloy "E".	2 in. outside dia. \times 0.12 in. wall.	0.29	0.48	0.20

Although the grain-size for the pure lead appeared very irregular, there was no sign of any marked grain-zoning in either this or the alloy "E" pipe. Other metallographic features are described below. Each pipe was 200 yd. long, and the major part of this was subjected to constant-pressure tests. The samples used for the constant-stress tests varied in length from 5 to $2\frac{1}{2}$ yd. The samples were made up in the manner standard for the pressure tests: brass caps were plumbed on to the ends, and the pipe was then taped for a few inches near each plumb. Pressures were read off by means of Bourdon gauges. For the higher pressures, water was used, obtained from a special gas pressure accumulator and connected to the test sample through a reducing valve regulator. For lower pressures, nitrogen gas from a cylinder was used and passed in through a Schrader valve, the necessary pressure adjustments being made by releasing gas as required. Measurements of diameter were made by means of diameter tape. The maximum accuracy in pressure measurements was to 0.25 lb./in.², and in diameter measurements to 0.005 in. In the original tests, using 5-yd. samples, diameter measurements were taken at several points along the pipe, and the pressure adjustments based on the mean. In later tests, using shorter samples, diameter measurements were taken at one point at the centre of the sample only, partly to save material, partly to simplify technique, and partly because it seemed logical to base readings on one locality (see below). The required pressures were obtained from a table of P against d_0 for each sample, at pressure intervals of about $\frac{1}{4}$ – $\frac{1}{2}$ lb./in.², the adjustments of pressure being made when the diameter had reached the value corresponding to each successive pressure reading on the table.

IV.—REVIEW OF THE RESULTS OF CONSTANT-PRESSURE TESTS.

Before giving the results of the constant-stress tests, a brief review of some outstanding results from the ordinary constant-pressure tests might be of value. In these tests the mean diameter of pipe samples (usually 5 yd. in length) is determined, the pressure, corresponding to a desired original hoop stress, being kept constant and diameter-tape measurements being taken at intervals until the sample bursts. Some of these tests, at very slow creep rates, have been in progress for more than 10,000 hr. The type of creep curve obtained for these tests is similar to that for creep of wire, strip, &c., under constant load. An initial rapid extension (stage 1) is followed by a slower extension, in most cases at a nearly constant rate (the so-called stage-2 creep), the final rapid (stage 3) extension taking place locally at the point of failure.

Typical creep curves, corresponding to a similar range of stress as used for the constant-stress tests are shown in Fig. 1 (pure lead) and Fig. 2 (alloy "E"). In general, there is little sign of stage-3 creep on the curves, as in calculation of the final extension the reading at the point of burst is omitted.

Briefly, the most important result obtained from these tests is that, whereas the extension obtained with alloy "E" is considerably greater at high rates of extension than with pure lead, a marked decrease occurs with alloy "E" when the rate of extension is less than 0.1%/hr. This decrease was not observed with pure lead. This effect is shown in Fig. 3, which shows the percentage extension obtained at burst (the mean along the sample length, omitting the value at the actual point of burst), against the mean rate of extension (final mean percentage extension divided by the time of test). The dotted line shows the average curve obtained for alloy "E", based on a large number of results of tests on material produced both on the ram-type and continuous screw-type presses. It will be seen that, whereas the mean extension finally obtained at extension rates greater than 0.1%/hr. is of the order of 16% (being considerably greater, 20% or more, when measured on the diameter at the point of fracture), the value obtained at extension rates of the order of 0.0001%/hr. is only about 2%. In the latter case, also, the extension at the point of fracture is no greater than this mean value.

The same diagram shows also the results of various individual tests, over the same range of extension rates, on (nominally) pure-lead samples from pipe produced on the continuous machine. The variability of extension both in the same pipe and in pipe from different sources and of different

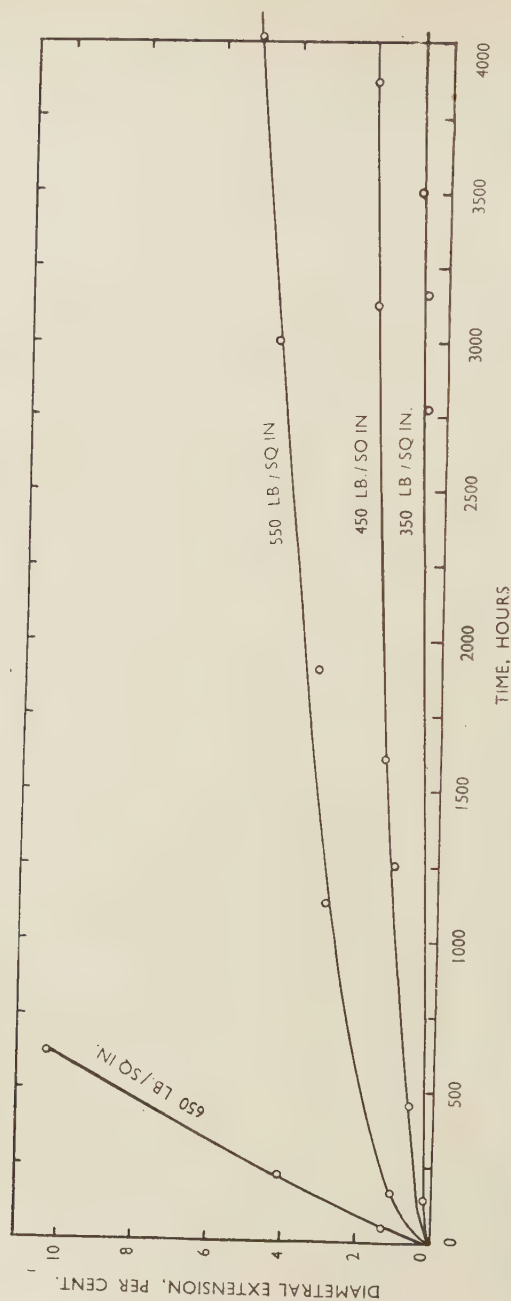


FIG. 1.—Constant-Pressure Creep Curves for Pure-Lead Pipe. Nominal hoop stresses in pounds per square inch shown beside curves.

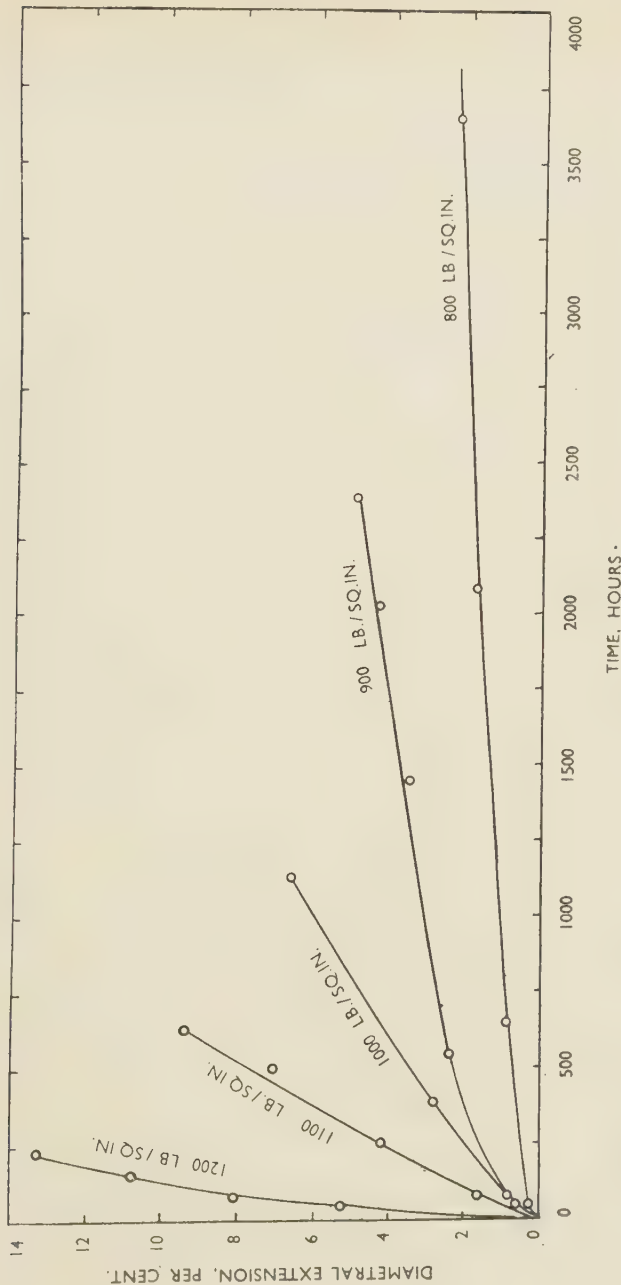


Fig. 2.—Constant-Pressure Creep Curves for Alloy "E". Nominal hoop stresses in pounds per square inch shown beside curves.

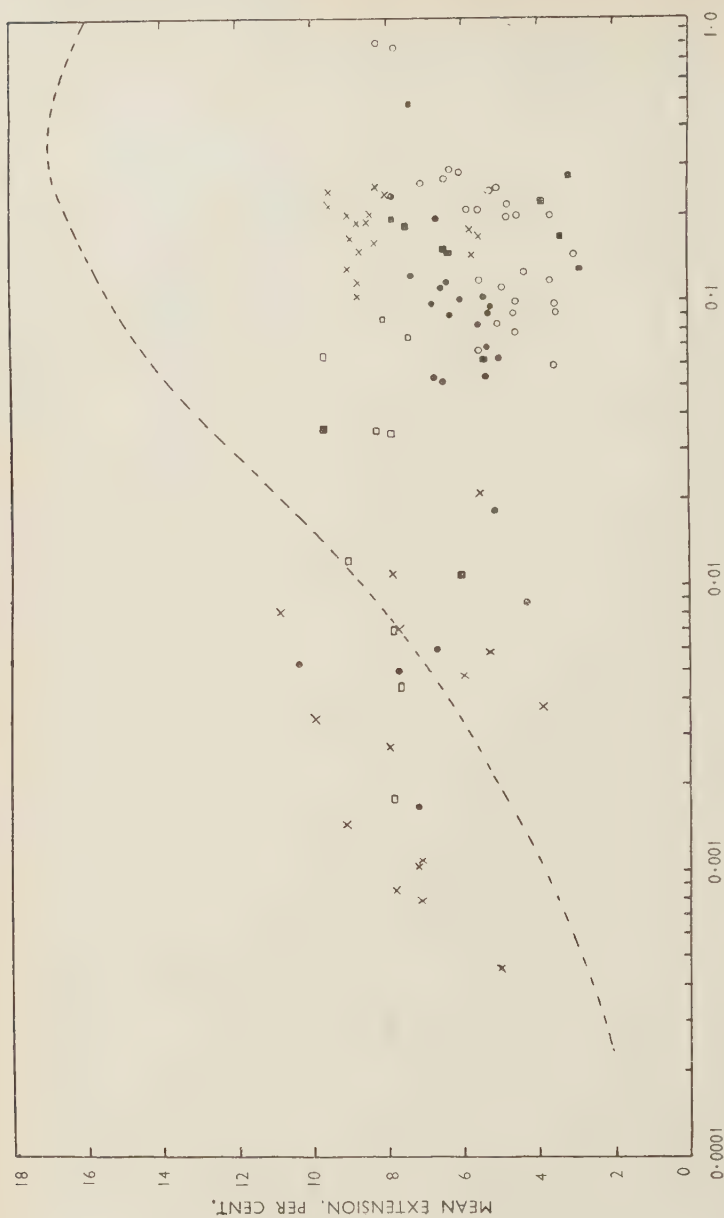


FIG. 3.—Diametral Extension of Lead Pipe. Various points show pure lead from different sources. Dotted line = alloy "E".

sizes is a feature of tests on pure lead. To some extent, the extension value appears to improve as the size of the pipe decreases and the grain-size decreases; these two conditions often go together. Apart from this, certain statistical conditions appear to obtain, but it is not proposed to deal fully with this aspect of the results here. The overall range of extension is of the order of 4–11%, and this range does not alter as the extension rate is diminished; that is to say, pure lead does not exhibit a diminishing "extensibility characteristic". Moreover, the extension at the point of break remains greater than the mean extension. There are also very significant differences in the type of fracture, which are discussed below. These results have been confirmed by a large number of tests carried out in the Technical Control Laboratories on both continuous-machine and ram-press pipe from a number of sources.

It was with the object of elucidating the reasons for this marked difference in behaviour between pure lead and alloy "E" pipe that the present work on constant-stress tests was undertaken.

V.—THE ANDRADE CREEP EQUATION.

The Andrade equation for creep at constant applied stress is:

$$l_t = l_0(1 + \beta t^{1/3})e^{kt},$$

where l_t = length of specimen after time t ; l_0 = a constant which in the present tests can be taken as the original length of the specimen; β = a constant associated with the so-called "grain-rotation" type of flow. This type of flow is often called transient flow or β -flow and appears to be dependent on a rotary motion, occurring within the crystal grains or, more accurately, grain fragments. It is considered to result essentially from the process of glide-plane slip (see below); and k = a constant, representing a type of flow considered to differ (perhaps fundamentally) from β -flow.

Andrade has applied this equation to the creep of lead wires, the load being applied by means of a weight of hyperbolic section partially immersed in water; the stress is kept constant by the automatic decrease in load as extension proceeds. A different device (balance beam with guide plates) has been used by Andrade and Chalmers,²⁷ and a further one (hyperbolic pulley) by Pearson.⁴ In applying this equation to pipe, it must be realized that the stress system is different from that of a wire or strip under tensile load. In the case of pipe, there is a longitudinal tensile stress parallel to the pipe axis and equal to half the tensile hoop stress; the other principal stress is a radial

compressive stress in the wall of the pipe. The difference in stress system causes a difference in the process of creep flow, but there is no reason to expect the fundamental mechanism of creep to be altered, and it is believed quite legitimate to apply the Andrade equation. This is confirmed by the good agreement obtained in most tests between the experimental and theoretical creep curves. The actual and relative values of the constants β and k may, however, be expected to be different for wire and pipe at equal applied tensile and hoop stresses.

There is another difference between pipe and wire constant-stress tests, in that in wire or strip the rapid stage-3 extension, which appears to be essentially due to stress concentration, is eliminated. This is not always true in the case of pipe, owing to the localized nature of the stage-3 creep. This does not, however, affect the calculation of β and k , as stage-3 extension does not enter into the calculation, and does not usually appear on the pipe creep curves.

It should also be borne in mind that the extension values given are based on the outside diameter of the pipe, and the extensions on the outside and inside of the pipe wall may be expected to differ somewhat. In pipe of the sizes tested, however, this difference is small enough to be neglected.

In applying the Andrade equation to pipe tests, it may be written in the form :

$$\epsilon = (1 + \beta t^{1/3})e^{kt} - 1,$$

where ϵ = diametric extension per unit diameter = peripheral extension per unit periphery = $\frac{l_t - l_0}{l_0}$.

It is not difficult to calculate β and k from the creep curves. For small values of kt the approximation $e^{kt} = 1 + kt$ leads to the approximate relationship : $\epsilon = \beta t^{1/3} + kt$, from which the calculation is simple. More accurate values can be obtained by successive approximations or direct curve fitting, &c.

VI.—RESULTS OF THE CONSTANT-STRESS TESTS.

Figs. 4, 5, 6, and 7 show extension-time curves for the results on pure lead. Some of these are completed tests, i.e. taken to bursting, some have been discontinued, and some have been continued for considerably longer times than those shown. There is, however, a limit to the useful time of a test, so far as constancy of stress is concerned, in that, at very slow creep rates, the pressure adjustments become too infrequent.

The full lines show the experimental curves, and the dotted lines the theoretical curves as obtained with the calculated Andrade constants.

It will be seen that in many cases the agreement with the experimental curves is quite remarkably good. In other cases there are discrepancies, particularly when longer times are reached. In the shorter-time

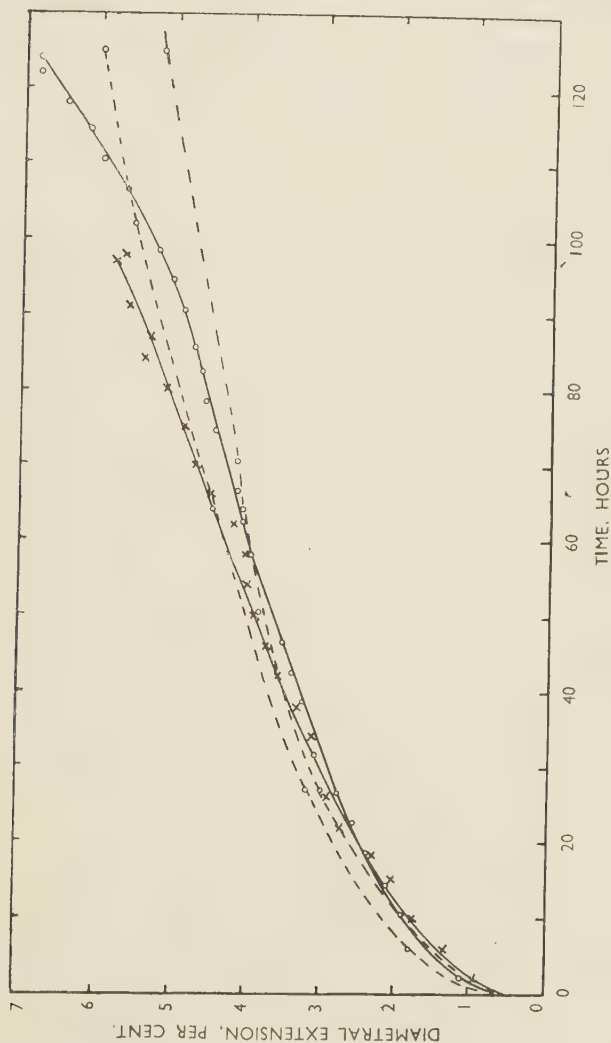


FIG. 4.—Constant Hoop-Stress Tests on Pure-Lead Pipe. Hoop stress = 800 lb./in.²
Full lines = experimental curves.
Dotted lines = Andrade curves.

tests at the higher stresses, some discrepancies occur which can be traced to the incidence of stage-3 creep at one or more localities near the points of measurement. Fig. 7 shows a group of curves from

results taken at one measuring point only, as described above. There is fair consistency between the lower-stress curves and those derived from mean measurements at a number of points, shown in Figs. 5 and 6, but the curve at 650 lb./in.² hoop stress is considerably different from the corresponding curve of Fig. 6. This is doubtless partly due to the fact that it is based on a sample from a different pipe, and of a different

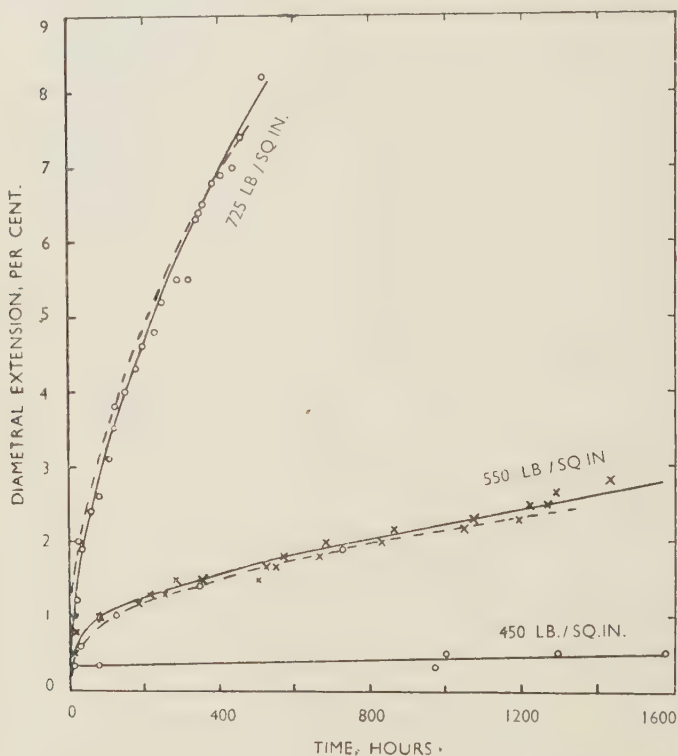


FIG. 5.—Constant Hoop-Stress Tests on Pure-Lead Pipe.

size. On the whole, the results were more consistent for the 3-in.-dia. pipe than for the 2-in.-dia. pipe; this was not unexpected and is probably due to the greater errors induced by dimensional factors in the smaller pipe.

It should perhaps be mentioned that considerable differences of creep rate at given pressures can occur in constant-pressure tests, both at different localities on a given (say 5-yd.) sample, and more particularly on different samples from the same extrusion. It is

hoped eventually to carry out an analysis of the considerable amount of data now accumulated, in order to determine possible causes of these variations. Insufficient metallographic differences were observed to make any structural correlation generally possible. Local variations in wall thickness may be partly responsible, although the eccentricity of these pipes was within the fairly close limits (± 0.005 in. variation

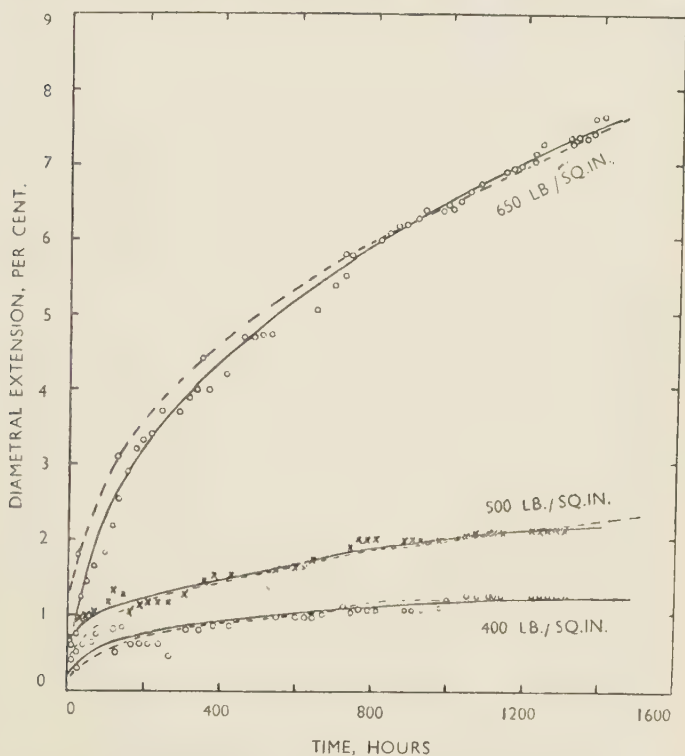


FIG. 6.—Constant Hoop-Stress Tests on Pure-Lead Pipe.

on mean wall thickness) specified. Measurements taken at the end of tests, however, showed that wall-thickness differences sometimes become accentuated to some extent. There are probably other factors, such as local hardness variations, also responsible. It is worthy of note that creep-rate differences tend to be accentuated in constant-pressure tests, owing to the stress-concentration factor; and in this respect, constant-stress tests have some practical advantage, even for routine tests. The correction is, however, insufficient if local creep-rate variations are of a high order.

On the whole, the conclusion is that, considering difficulties in technique and possible sources of error, the results are in reasonably good agreement with the Andrade equation.

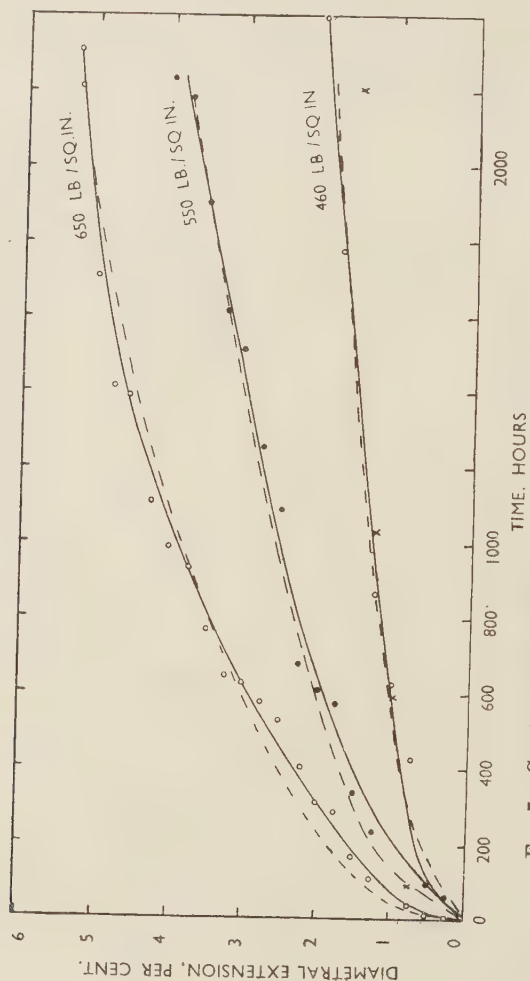


Fig. 7.—Constant-Stress Tests on Pure Lead. Single point measurements.

Results on alloy "E" are shown in Figs. 8, 9, 10, and 11. The curves in Fig. 11 are from single-point measurements, and the others are based on mean measurements. The dotted curves are, as before, from the theoretical Andrade relationship. The agreement is again, in many cases, quite striking; but on the whole, the results

are not as consistent as those for pure lead. This was expected, as general creep-rate variations and local variations were usually greater for alloy "E" than for pure lead.

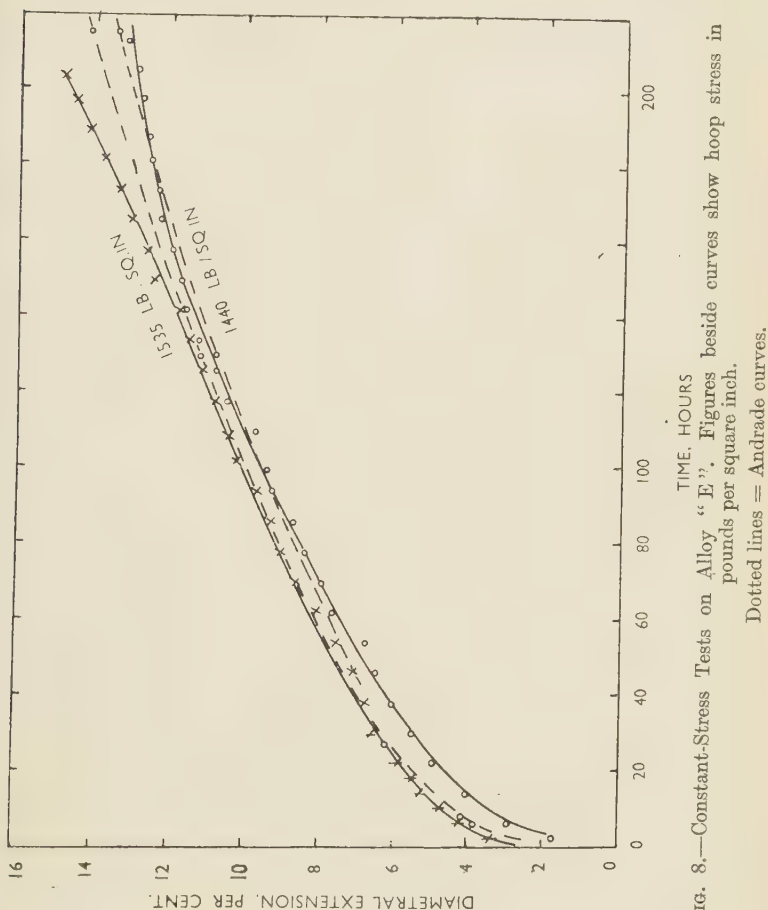
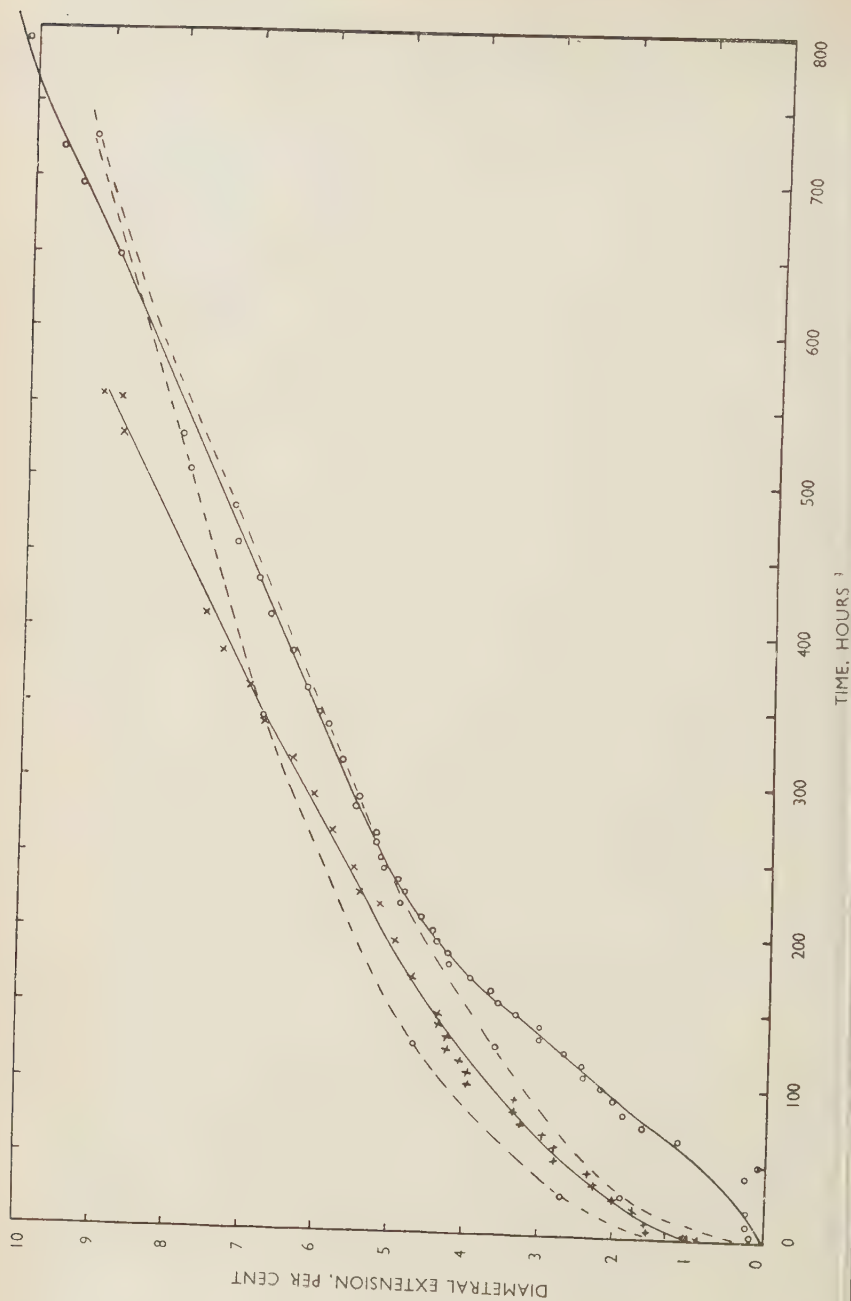


Fig. 8.—Constant-Stress Tests on Alloy "E". Figures beside curves show hoop stress in pounds per square inch. Dotted lines = Andrade curves.

One of the curves at 1200-lb./in.² hoop stress is inconsistent at the start of test, and no really satisfactory reason for this was apparent. The first part of the test period is naturally somewhat difficult to control fully; but in view of the good results obtained with other tests, this did not appear to be sufficient explanation. Some inconsistencies have been found in constant-pressure tests carried out at stresses of this order, particularly at about the corresponding range of creep



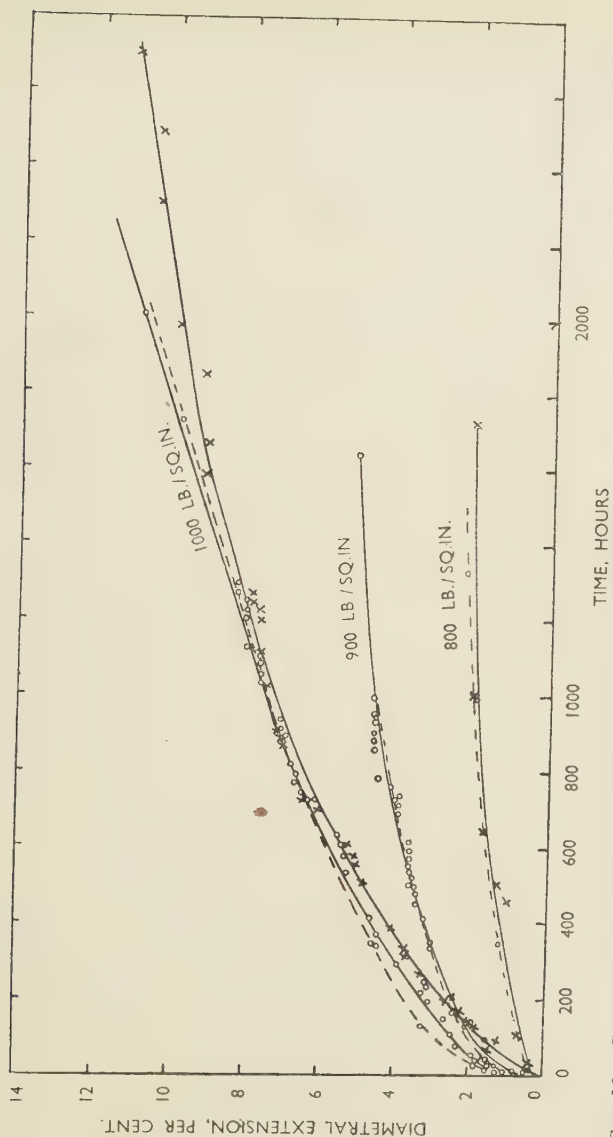


FIG. 10.—Constant Hoop-Stress Tests on Alloy "E". Figures against curves show stress in pounds per square inch. Dotted lines = Andrade curves.

rates. The possibility, therefore, arises of some disturbing action having taken place in the alloy, particularly noticeable at creep rates of this order; but there are insufficient data for confirmation of this. This point is dealt with again below.

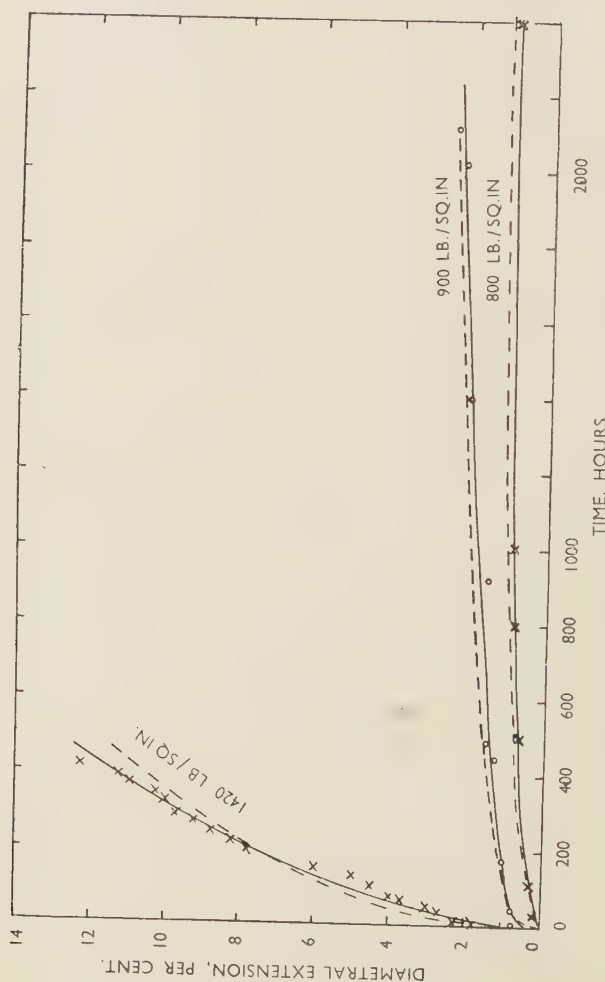


Fig. 11.—Constant-Stress Tests on Alloy "E". Single point measurements.

It is worthy of note that checks on the hoop stress carried out at the end of constant-stress tests usually gave good results. Thus, for alloy "E", at 1535 lb./in.² hoop stress, the final average wall thickness was 0.105 in. (variation 0.1002–0.1107 in.), the mean outside

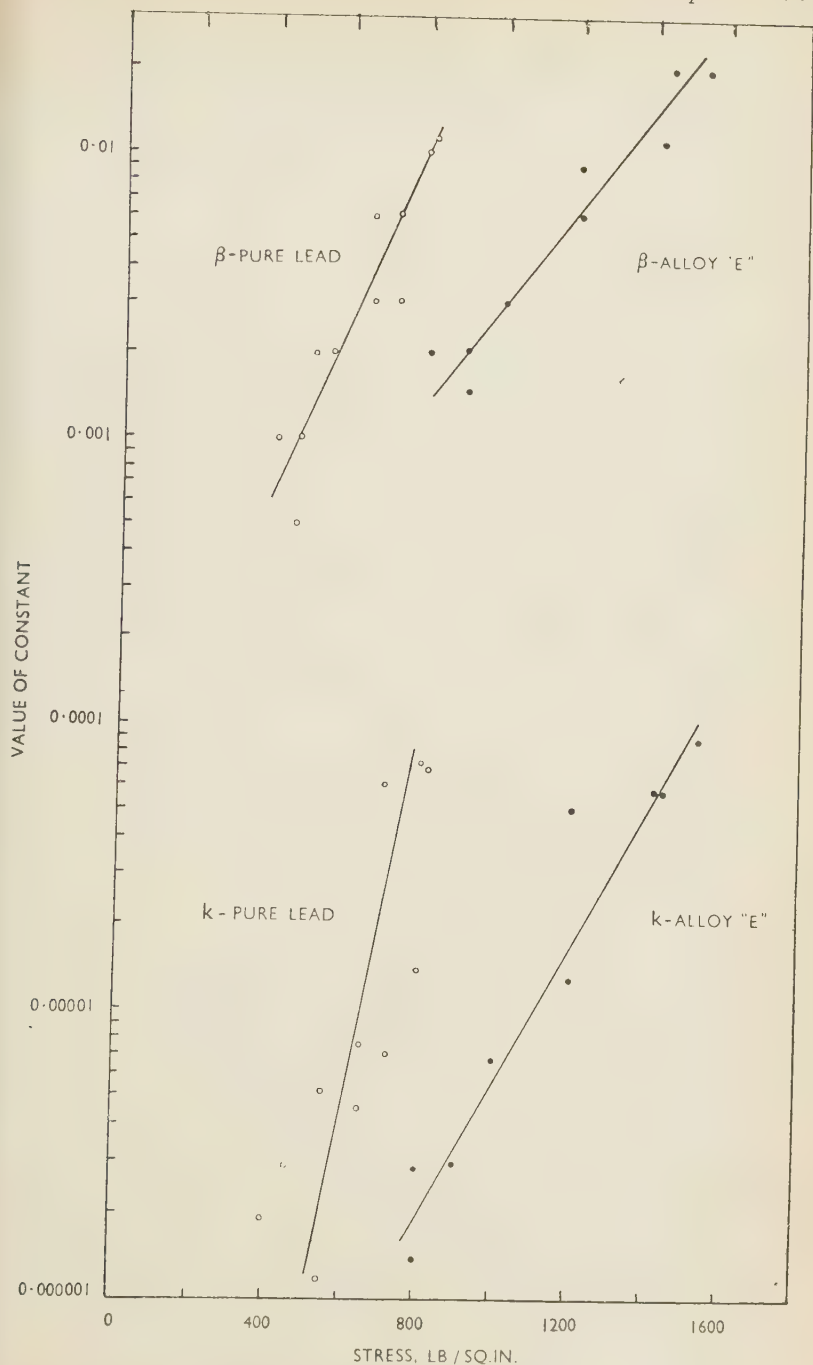


FIG. 12.—Andrade Creep Constants.

diameter being then 3.45 in. (variation 3.405–3.520 in.), and the final pressure 96.75 lb./in.² This corresponds to a hoop stress of 1500 lb./in.² Similar figures for another alloy "E" sample at 1400 lb./in.² hoop stress were :

Mean wall after test :	0.106 in. (variation 0.0955–0.1175 in.).
Mean external diameter after test :	3.406 in. (variation 3.365–3.440 in.).
Final pressure :	885 lb./in. ²
Calculated hoop stress:	1330 lb./in. ²

These tests were carried out in the least-favourable conditions (high final extensions and maximum diametral variations), so it can be concluded that an error of more than 5% in the hoop stress is unlikely. For most tests, the error was actually well within 5%, and in many cases within 2%, as determined in this way.

The derived β and k values are shown in Fig. 12, plotted semi-logarithmically against the hoop stress.

There are some inconsistencies, but there seems little doubt about the order of values, and that these are much the same, over the different range of stresses used, for pure lead and alloy "E". The points appear to lie approximately on straight lines, suggesting an exponential relationship with stress, possibly of the type :

$$\beta = A(e^{af} - 1).$$

The significance of the constants A and a is not obvious, and no attempts have been made to determine them. Actually, Andrade has shown that β reaches a maximum value at a certain stress, which must be considerably higher than any used for the present tests.

VII.—DISCUSSION OF RESULTS.

The range of hoop stresses over which these tests have been carried out, namely 800–400 lb./in.² for the lead, and 1400–800 lb./in.² for alloy "E", corresponds to a similar range of creep rates, namely from about 0.1–0.0001%/hr., or less. Over this range, the β value is, for both materials, considerably in excess of the k value. From the similarity of the β curves and k curves respectively, the important conclusion may be reached that, for similar creep rates, there would appear to be little fundamental difference between the mechanism of creep flow in the lead and alloy "E" pipes. In fact, the curves for alloy "E" have a smaller slope than the corresponding curves for lead. It is possible that this is connected with the difference in

grain-size, that for the alloy "E" pipe being considerably smaller than that for pure lead, as shown in Table I. For the same stress, as will be seen for the values at 800 lb./in.², the values of β and k are much smaller for alloy "E" than for pure lead; this is due to the effect of the alloy constituents.

Another conclusion of importance is that, as the applied stress diminishes, the proportion of β -flow in both materials increases, since the value of k falls to a very small figure, while the value of β is still quite appreciable at the lowest stresses at which the tests have been carried out. It would seem possible, therefore, that at small creep rates the resistance of the pipes to failure is profoundly influenced by their resistance to the effects of grain rotation.

As has been described above, the behaviour of the lead and alloy "E" pipes, in particular as regards the extension reached at different creep rates before failure, is greatly different. This difference may now be summarized as follows :

(1) At high creep rates (or strain rates), there is a considerable amount of both β -flow and k -flow, and, under these conditions of combined flow, the extension reached by the alloy "E" pipe before failure is considerably greater than that reached by the lead pipe.

(2) At small creep rates, or, more accurately, small applied stresses, the flow is chiefly or largely of the β -type. Under such conditions, while the extension of the lead pipe is (so far as tests have been able to show at present) not greatly influenced, there is a very marked decrease in the extension of the alloy "E" pipe.

(3) Another important difference is in the mode of fracture of the two materials. For alloy "E", the fracture is of an intercrystalline nature (so-called "creep-type") at all but the highest strain rates, when it approaches the "long-break", or so-called "knife-edge", type, which appears to be chiefly transcrystalline. For lead, the fracture is of the latter type throughout, although there is some intercrystalline cracking at small creep rates.

For alloy "E", at small creep rates, there is little if any difference between diametral extension at the locality of failure and the mean diametral extension along the sample length. For the lead, the extension at the point of fracture is, in general, greater than that remote from the fracture. This indicates that stage-3 creep disappears from the alloy "E", while persisting in the lead.

A hypothesis to account for these differences, which are of great practical importance, is given below, but first it is necessary to deal somewhat more fully with some fundamental aspects of creep flow.

VIII.—FUNDAMENTALS OF CREEP FLOW.

A large amount of physical and mathematical thought has been expended by many workers during recent years in an endeavour to understand the nature of the process of creep in both metallic and non-metallic materials, to explain the many complicated phenomena which have been observed, and to express the results of the measurements of creep rate under different conditions, in precise terms. It may be stated at once that no single theory has yet proved adequate, and that the simplifications involved in obtaining the mathematical relationships have raised objections, in many cases, to their application. In particular, none of the theories or mathematical relationships has yet succeeded in forecasting with sufficient precision the incidence of failure.

In the purely theoretical field, the work of Eyring^{5, 15} on the application of activation-energy concepts, applied to metals by Kauzmann⁶ and others, has had considerable influence. In the present paper, however, a simpler approach to an understanding of the mechanism of creep flow is attempted.

Many creep-rate equations, such as the Andrade equation, have included terms containing fractional powers or logarithms of time.⁸ This type of term is associated with what has been called "transient flow", the fractional power or logarithm indicating the transient nature of the flow, in contrast to the linear time term associated with "permanent" flow. Thus, in the approximate form of the Andrade equation for small creep rates:

$$\epsilon = \beta t^{1/3} + kt$$

the first term represents "transient" flow and the second "permanent" flow. For many metals, particularly in tests at constant load, the transient term does not appear explicitly in the equation for stage-2 creep. The equation $\epsilon = \epsilon_0 + vt$, where ϵ_0 represents a constant connected with an initial rapid deformation, would seem to apply fairly exactly. There appears to be no really satisfactory explanation why such a simple proportionality law should apply to these conditions; it is certainly by no means universally applicable, even to metals. Logarithmic expressions have often been applied, particularly to plastics &c.^{7, 8} For example: $\epsilon = \epsilon_0 + at + b \log t$. Here again, it is possible to associate the logarithmic time term with transient flow, although other explanations have been offered.

Andrade called the transient-flow term in his equation " β -flow", and it has been connected with grain rotation (see below). It may

be noted that the Andrade equation is probably not universally applicable, but has been found to apply very well to many metals, such as tin, lead, copper, aluminium, iron, and cadmium.^{2, 27}

IX.—GRAIN ROTATION.

The rotation of crystallites, accompanying deformation, is apparently quite a common phenomenon in the plastic flow of metals. Moore and his co-workers¹⁰ have demonstrated its occurrence in the creep of lead and its alloys by the distortion of microtome lines. The present author has observed grain rotation (or at least relative movements at grain boundaries) by the distortion of "flow lines" or "ghost lines" frequently present in extruded pipe. It can be shown mathematically¹¹ that any strain (other than so-called "pure strain") is associated with rotation as well as deformation, so the occurrence of rotation, accompanying deformation of crystals as the result of plastic flow or creep, is to be expected. Moore¹⁰ has explained the process, in the creep of lead, &c., as being the result of a torque, exerted by neighbouring grains at grain boundaries by the process of slip within the grains. Whatever its exact mechanism, there can be little doubt that the rotation process accompanies the ordinary process of slip within the grains, probably in a similar manner to the rotation of slip planes in a single-crystal specimen submitted to a tensile stress.

The measurements of Andrade and Chalmers,²⁷ of the changes in electrical resistance of polycrystalline wires of cadmium and tin undergoing creep at constant stress, have indicated that grain rotation of this nature is closely associated with β -flow, appearing to diminish as β -flow diminishes at any given applied stress. Confirmation of their results has been obtained in the X-ray studies of Gibbs and Ramlal.⁹ The work of Taylor,³¹ summarized, together with his own work and that of others, by Barrett,³² on the production of textures by plastic deformation, giving a detailed analysis of the grain rotations and slip processes concerned, is also of moment in this connection.

With regard to grain rotation, as associated with β -flow, it seems logical to deduce that its influence would show very markedly at grain boundaries. There is no reason to suppose any co-ordination between the rotary processes in adjacent, differently oriented grains, and hence considerable relative motion at grain boundaries would be expected. In other words, the development would be expected of what perhaps amount to large shear strains, and possibly also the occurrence of actual voids. The conclusion is therefore reached that grain rotation, as occurring in association with β -flow, is a factor which could lead to disruption at grain boundaries.

The flow associated with the k -constant of the constant-stress tests is not regarded by Andrade as merely being a modified continuation of the flow associated with the β -constant. He regards it as of a fundamentally different nature, occurring, as the equation indicates, together with the β -flow, and continuing when the β -flow falls off. It seems extremely likely, however, that the process of glide-plane slip is an integral part of k -flow, although possibly differing in important respects (such as numbers, spacing, or even orientations, &c., of glide planes) from the slip associated with β -flow. According to the theory at present most generally accepted, slip takes place by the travel of dislocations, of atomic dimensions along the slip planes.¹² Mott¹³ considers that grain boundaries are the most likely sources of such dislocations, and while the precise mechanism of the formation and motion (at grain-boundary regions, at least) of dislocations is far from being fully understood, it is reasonable to deduce that the grain boundaries play an important part in k -flow.

Several authors, for example Pearson,⁴ have put forward the view that k -flow or "viscous flow" is chiefly a grain-boundary process, at least at small creep rates. Zener¹⁹ has described a process of flow (applicable to creep) which depends on the assumption that the grain boundaries behave as though a viscous film, a few atoms thick, were present; although apparently the presence of such a film as a definite and separate entity is not claimed. Kê²⁹ has studied "Zener anelasticity" experimentally in aluminium and magnesium, and claims that his results are in favour of Zener's theories. Kê's studies included constant-stress creep tests on aluminium wires under torsion; the conditions, however, differed considerably from those with which the present work on lead is concerned.

On the whole, it is tempting to regard k -flow as in itself possibly made up of flow processes both within the crystallites (through perhaps the ordinary, or even a modified, mechanism of slip) and at the grain boundaries, perhaps by processes not dissimilar to Zener flow. The apparent simplicity of k -flow, therefore (as regards its time dependence), is perhaps misleading. It does not seem illogical to assume that k -flow or "viscous" flow would help in filling in any grain-boundary discontinuities or voids and therefore oppose the disrupting effect of grain rotation. This gives a simple picture of β -flow as a possible cause of grain-boundary disruption, and k -flow as a grain-boundary "mender". Such a picture of the creep and fracture processes is obviously highly simplified, but it enables an explanation to be offered for the extension results obtained with alloy "E" and lead pipes.

X.—STRESS RELAXATION.

If deformation be applied to a sample of material under conditions in which it will undergo creep, and the strain be kept constant, relaxation of stress occurs according to some such equation as :

$$\sigma_t = \sigma_0 - \beta \log (t + 1).$$

Reference should also be made to the equation of Trouton and Rankine,¹⁴ of similar type.

σ represents stress and β a constant which is associated with transient flow probably of very similar nature to the β -flow described above, although the precise relationship between the β of the Andrade equation and the β (or β/σ_0) of the stress-relaxation equation needs to be established. They are obviously not the same. This process of stress relaxation seems likely to be of some importance in connection with sheath for gas pressure cable. In particular, it is suggested that it could readily lead to intercrystalline failure,¹⁹ a point which is dealt with again below.

Before leaving the general discussion of creep theory, mention should be made of the work of Orowan,^{3, 16} Burgers,¹⁷ Dushman,¹⁸ &c., dealing with the application of activation-energy concepts; that of McAdam and his co-workers,²⁰ dealing with a conception of "technical cohesion limits"; as well as the recent Royal Society discussion,³ and A.I.M.M.E. symposium,²¹ to mention but a few sources of often contrasting theories and concepts. Recent X-ray-examination results reported by Wood and Tapsell²² may also be mentioned. They are apparently somewhat different from the results of Gibbs and Ramlal,⁹ described above, although not necessarily in opposition to the grain-rotation theory. Critical co-ordination of these many existing theories and complex observations would be a task of the greatest difficulty, but would undoubtedly be of real benefit.

XI.—A THEORY OF THE CREEP BEHAVIOUR OF LEAD AND ALLOY "E" PIPE.

Based on the simple creep mechanism propounded above, a theory (which, indeed, claims to be little more than a working hypothesis) to explain the different behaviour of pure lead and alloy "E", can now be offered. A basic assumption in this theory is that pure lead is able to withstand the effects of grain rotation better than the alloy. Such an assumption is not unreasonable; it is not difficult to conceive of local flow or "recovery" effects at grain boundaries taking place more readily in pure lead than in the harder material. On this assump-

tion, the difference between pure lead and alloy "E" can be accounted for as follows:

(1) At high creep rates or extension rates, there is a considerable amount of "viscous" flow (of glide-plane slip type), as well as grain rotation, in both materials. According to the theory given above, this allows alloy "E" to withstand considerable extension before failure occurs. The disrupting effect of grain rotation is overcome by the "viscous" flow of fresh material to the grain-boundary voids. The greater extension of the alloy, as compared with pure lead, would then appear to be quite logically accounted for by the retardation, in the harder material, of the start of the final stage-3 local extension, which precedes failure.

(2) As the creep rate (or, more accurately, the applied stress) diminishes, there is an increase in the proportion of β -flow for both materials. Grain-boundary disruption takes place at an earlier stage in alloy "E", resulting in a decrease in extension at failure with decreased creep rate, and in an eventual disappearance of stage-3 creep. Pure lead, on the other hand, can withstand more grain rotation, the stage 3 is still reached, the fracture remains of transcrystalline type, and there is no substantial alteration in extensibility. At very slow creep rates, there is probably an increasing tendency for intercrystalline failure, but the extension will probably be greater than corresponding values for alloy "E". The values for the extension of the pure-lead pipes at *very* slow creep rates are not yet known, however; the tests are still in progress, and it will be some time before they are completed.

It must be emphasized that this description of the process is not intended to be more than a simplified picture. It is quite probable that there are other factors concerned, and one possibility may briefly be mentioned. It is known that ageing in lead-antimony alloys can be accelerated by cold work. Alloy "E" is a solid solution, but is rather near the limit of the solid-solubility range. The separation of antimony (or perhaps tin-antimony compound) at grain boundaries as a result of creep is within the bounds of possibility, although metallographic examination has failed to disclose any signs of this. On the other hand, differential etching effects, associated with the chemical deposition of antimony which takes place with the usual etching reagents, have indicated the possibility of some sort of change resulting from creep, particularly at grain boundaries. Such effects, however, appear more clearly in samples from the rapid tests than in those from the slow creep tests.

A further point which must be emphasized is that there is no

suggestion that a certain degree of β -flow (corresponding to 2% extension, say) will, in the absence of k -flow, inevitably lead to *immediate* failure in an alloy of the type of alloy "E". Intercrystalline failure at small extensions (or, more accurately, small creep rates) appears to be a very gradual process, quite possibly, in fact, taking place by distinct steps. The slowness of the development of failure has been well demonstrated by certain experiments carried out by the author in the Research Laboratories. These tests have also demonstrated in a striking way the very large relative movements at grain boundaries which occur both in pure lead and alloy "E". Analysis of the Andrade equation shows that the major portion of the β -flow takes place in the earlier stages of creep (hence, indeed, the term "transient"). The suggestion here is that if this is followed (or accompanied) by sufficient k -flow, failure does not occur until a considerably greater extension is reached; but if the amount of k -flow is too small, as at slow creep rates, failure gradually develops. In other words, β -flow puts the grain boundaries in some sort of "state of strain" or even of incipient failure, which is relieved by k -flow, but gradually leads to complete failure if k -flow is insufficient, especially in alloy "E" and other alloys which behave similarly.

The idea of slow disruption at boundaries (or on slip planes) leading to failure in creep is not new. Hanson and Wheeler,²⁸ from their metallographic and density studies of aluminium specimens undergoing creep (the appearance of slip lines and grain boundaries on polished surfaces was studied in particular), put forward the view that some such process was occurring. Their studies also afforded some evidence of relative grain movements at boundaries.

It is of interest to note that, if the reality of slow disruption were acceptable and "disruption rate" were represented by some definite value (V_1 say, with V_2 representing the creep rate), then (assuming for simplicity a linear relationship with time) the extension obtained at failure could be conceived as depending on some such relation as V_2/V_1 , the determination of which again involves, basically, an analysis of the relationship between the creep process and the fracture process.

In developing any complete theory of creep fracture, the possibility must be considered that there may be differences in the effects of k -flow itself in pure lead and the alloy. It is conceivable that relative grain movements occur during k -flow, possibly to a different extent, or having different effects, in the different materials.

Recently, Mott³⁰ has put forward the theory that the decrease of extension at fracture with decrease of applied stress is connected with the relative effects of Zener flow (i.e. viscous boundary flow) and the

ease of deformation or yield of the grains themselves. Viscous movement along the grain boundaries can, by mutual interference or "interlocking" effects between adjacent grains, produce large disruptive forces. At high, externally-applied stresses, these are relieved by the deformation of the grains themselves. At small applied stresses, the effect of Zener flow is much greater in proportion. The rate of viscous flow has a linear dependence upon stress, the rate of grain deformation decreasing exponentially with stress. More convincing evidence of the reality of Zener flow, and in particular its linear dependence upon stress in a polycrystalline mass, would appear to be desirable. The theory also implies that "diminishing extensibility" in creep would be universal, and there does not appear to be any good reason for supposing that this is so, even in the absence of recrystallization or other complicating factors.

On the whole, the general picture of some sort of flow (or deformation process) producing or leading to disruption, in particular at grain boundaries, and some other sort of flow (or deformation process) having a compensating, relieving, or "mending" effect, would appear to give a sound basis for the understanding of the fracture process in creep. Possible changes in available slip systems necessary for grain conformation may also be of importance in both β - and k -flow.

XII.—GRAIN-SIZE.

The importance of the effect of grain-size in the failure of lead pipe, &c., under creep has been pointed out by McKeown.²⁴ Grain refinement would appear to be of benefit in avoiding intercrystalline cracking. There are no data on the precise influence, on β and k values, of grain-size, but it seems reasonable to postulate that the smaller the grain-size, the more readily can mutual accommodation of adjacent grains, under the influence of grain rotation, take place. It is quite likely also that the relative values of β and k are favourably altered. Some results obtained by the author on rolled alloy "E" strips submitted to static load tests are not without significance in this respect. "Knife-edged" or "long-break" fractures were obtained at stresses which give short breaks of intercrystalline type with strip specimens taken from pipe. The effect would appear to be due to grain refinement.

The decrease in extension with decreased extension rate obtained with alloy "E" pipe, has been observed in a variety of other alloys tested. Other workers have reported decreases of this kind, but few published results have shown effects as marked as those obtained

in the author's laboratories. On the other hand, the tests have not usually been so comprehensive. Some tests of Gohn, Arnold, and Bouton,²⁵ on extruded strip extended over very long periods, are of great interest. Some of their results apparently contradict pipe-test results obtained on similar materials in the laboratories of British Insulated Callender's Cables, Ltd. It is possible that grain-size differences may account for these contradictions. Gohn, Arnold, and Bouton found considerable evidence of recrystallization and ageing phenomena in some of their alloys, which may have influenced matters.

XIII.—PRESSURE-CABLE SHEATH.

Lead sheath for pressure cables with internal gas pressure of the order of 200 lb./in.² must be reinforced. On subjection to pressure there is an initial extension of the sheath against the outer reinforcement, and tests have indicated that this may be followed, with the usual systems of reinforcement by copper or steel tapes, &c., by very slow creep. Sheath for such cables must, therefore, have good extensibility and withstand the effects of slow creep. These may be aggravated by certain extrusion defects, such as the presence of inclusions, charge tongues,²⁶ zones of different grain-size, &c. It is also possible that stress-relaxation effects, due to the relief of internal stresses, may be of importance. On the above-mentioned theory, it becomes possible that the criterion for a suitable sheath is that it should be made of an alloy which will withstand β -flow, and in a condition most favourable for withstanding the effects of β -flow. In this latter connection, fineness and uniformity of grain-size are factors to be borne in mind.

XIV.—CONCLUSIONS.

Briefly, these are the most important conclusions to be reached :

(1) The results of the constant-hoop-stress tests indicate that, within a similar range of creep rates, the fundamental mechanism of creep flow in the pure-lead and alloy "E" pipes under test is not greatly different. In particular, the corresponding Andrade β and k constants are of a similar order and decrease in a similar way with decreased applied stress. The possibility of certain secondary effects in alloy "E" cannot be excluded, but there is insufficient evidence to show their precise nature or importance.

(2) For both materials, the indications are that, as the applied stress diminishes, the k constant decreases to a very small amount, while the β constant still has an appreciable value, i.e. the proportion of β -flow markedly increases.

(3) The decrease in extension which the alloy "E" pipe is able to withstand before failure, and the increased liability to inter-crystalline fracture, with decreased mean creep rate may be connected with the effects of β -flow when the amount of k -flow combined with (or following) it decreases. Pure lead, it is postulated, is better able to withstand these conditions than alloy "E".

(4) Failures in pressure-cable sheath of certain alloys may be connected with these effects and possibly also with effects, of a fundamentally similar nature, due to stress relaxation. These effects are, furthermore, aggravated by the presence of extrusion defects.

(5) The influence of grain-size needs further study, and grain refinement may prove to be of great importance, if carried out in a suitable manner, in increasing resistance to failure by β -flow, or whatever grain-boundary disrupting influence is concerned.

As a further point of some interest, it may be mentioned that both Andrade and Pearson found that alloys of the tin-lead eutectic type exhibit negligible β -flow ($\beta = 0$). Alloys of this type appear capable of very high extensions, and it is perhaps not too far-fetched to suppose that they would be ideal for pressure-cable sheath from this point of view. The cost, of course, is prohibitive, and other difficulties are likely, but such alloys open up interesting possibilities.

It is hardly necessary to emphasize finally that the theories given above are not claimed as complete. Further data will undoubtedly be necessary before a full explanation can be offered of the behaviour of alloy "E" and similar materials under prolonged creep. A crucial point, for instance, is that the effect of (rapid) β -flow, when combined with the (less-rapid) k -flow, corresponding to tests at the higher stresses, is, according to the theory developed above, less likely to cause failure (in alloy "E") at small extensions than the effect of slow β -flow, taking place over a long period with diminishing rapidity and combined with slow k -flow. While the evidence is in favour of this view, there is no direct proof. It is worth noting that the possible change in nature of k -flow at slow creep rates, mentioned above, may be of importance in this respect, while the precise manner of relative grain movements at boundaries needs further study.

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THE YOUNG'S MODULUS OF SOME ALUMINIUM ALLOYS.*

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WITH AN APPENDIX

THE MODULI OF ALUMINIUM ALLOYS IN TENSION AND COMPRESSION.

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SYNOPSIS.

Increased interest is being shown in the elastic properties of alloys for aircraft construction. Various binary, ternary, and complex aluminium-base alloys were prepared with a view to investigating the possibility of increasing their Young's modulus, E . All the five elements investigated were found to enhance the value of E , manganese having the greatest effect (about 0.34×10^6 lb./in.² for 1 wt.-%) and the elements beryllium, cobalt, nickel, and silicon having decreasing effect in that order. Beryllium and cobalt both have a greater effect in the presence of 12% silicon than in the respective binary alloys, but the effect of beryllium is smaller in alloys containing 2.5% copper. Attempts to relate these effects with the constitution of the alloys have met with only partial success. The values of E obtained on forged and heat-treated aluminium-nickel and aluminium-silicon alloys were in agreement with those given by corresponding alloys in the chill-cast condition.

On a complex aluminium alloy containing silicon 12, copper 2, nickel 2, manganese 1, magnesium 0.6, cobalt 0.2, and titanium 0.05% (with the exception of sand-cast and thin sheet materials, which can be explained), the value of E was the same, within experimental error, for materials in the cast, forged, and rolled conditions, both before and after heat-treatment. Complex alloys containing varying amounts of nickel and manganese gave higher values of E than those obtained by calculation from the previously-determined individual effects of the various elements.

Further developments of the work are discussed briefly, including the possibilities of developing alloys combining high strength with improved Young's modulus. Although no effort was directed towards this end, the indications are that, by suitable modification of composition, alloys can be produced having these tensile properties: 0.1% proof stress,

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21 tons/in.²; maximum tensile stress, 28 tons/in.²; elongation on $4\sqrt{A}$, 4%; Young's modulus, 12×10^6 lb./in.² It is considered that it will be possible, as a result of further work, to produce alloys with a Young's modulus greater than 12.0 without any marked lowering in these tensile properties.

The Appendix describes tests designed to determine whether, for aluminium alloys, the Young's modulus in tension is the same as the Young's modulus in compression. Two materials were used: one in the form of bar to British Standard Specification No. 6L1, and the other in the form of sheet to Specification D.T.D. 646A. The results indicated that, within the limits of experimental error, the moduli in tension and compression were the same.

I.—INTRODUCTION.

THE development of high-strength aluminium alloys, combined with certain trends in aircraft design, has led to a situation in which it is envisaged that the failure of a structure may be brought about by elastic instability of compression members before the tension members fail in tension. The onset of elastic instability is governed, other things being equal, by the Young's modulus of the material concerned; consequently, considerable attention has recently been paid to the possibility of improving aluminium alloys in this respect.

It is commonly believed that the Young's modulus of the alloys of a given basis metal is virtually constant, although it is recognized that preferred orientation may exert some influence in appropriate cases. As aluminium is one of the most isotropic metals from the elastic point of view, it follows that no useful increase in Young's modulus can in this case be expected from this cause. A discussion of these and other factors in relation to Young's modulus is given by Mack,¹ but he does not suggest any possibility of raising the Young's modulus of aluminium alloys; in fact, he states that "if it is desired to change E of a given material permanently, recourse must be had to purely mechanical means, such as the development of texture, combination or incorporation through suitable design of other substances with the desired material, or by increasing the porosity of the material itself".

It was nevertheless observed by the authors that certain experimental alloys^{2, 3} based on the aluminium-silicon eutectic gave abnormally high values of Young's modulus. Statistical analysis of the results available indicated that high Young's modulus was related to the composition; in particular, to the amounts of manganese and nickel in the alloys. Accordingly, it was decided to investigate the effects of various alloying elements, such as silicon, manganese, nickel, cobalt, and beryllium, on the Young's modulus. The work described in this paper was undertaken in order to arrive at an understanding of these effects and to provide the basic information that is required for the

development of alloys combining a high value of E with acceptable tensile and other properties.

Experimentally, it is more convenient to measure Young's modulus in tension rather than in compression, although it is in compression that the structural significance of Young's modulus is important. Experiments were therefore undertaken to determine whether there was any difference between the values in tension and in compression for a given material. This work and the results obtained are described by Grover, Munro, and Chalmers in the Appendix.

II.—EXPERIMENTAL PROCEDURE.

1. *Chill-Cast Bars.*

Small melts of about 3 lb. each of various binary, ternary, and quaternary alloys were prepared in a salamander crucible using an electric resistance furnace. The alloys had the following nominal compositions :

Binary Alloys :

(a) Aluminium-silicon with 5, 10, 15, 20, 25, and 30%, respectively, of silicon.

(b) Aluminium-beryllium with 1, 2, 3, and 4%, respectively, of beryllium.

(c) Aluminium-cobalt with 1, 2, 3, 5, and 8%, respectively, of cobalt.

(d) Aluminium-manganese with 1.5, 3, 5, 8, 10, and 15%, respectively, of manganese.

(e) Aluminium-nickel with 1, 3, 5, 8, 10, and 15%, respectively, of nickel.

Ternary Alloys : 12% of silicon and

(f) 1, 2, 3, and 3.5%, respectively, of beryllium.

(g) 1, 2, 3, 5, and 8%, respectively, of cobalt.

(h) 1, 3, 5, 8, and 10%, respectively, of manganese.

(i) 1, 3, 5, 8, 10, and 15%, respectively, of nickel.

2.5% of copper and

(j) 1, 2, 3, and 4%, respectively, of beryllium.

Quaternary Alloys : 2.5% of copper, 12% of silicon, and

(k) 1, 2, and 3%, respectively, of beryllium.

The melts were degassed and fluxed with from 0.2 to 0.4% of sodium boron fluoride; and chill-cast test bars, 1 in. dia. and 12 in. long, were

cast in iron moulds coated with graphite and having a wall thickness of about 0.5 in. The moulds were preheated to about 200° C. and were inclined at an angle of 45° to the vertical during pouring and remained in this position until the metal had completely solidified. The metal was poured slowly into the mould at from 700° to 920° C., depending on the alloying content. This method of pouring reduced the turbulence to a minimum and ensured a steady flow of metal, resulting in sound bars with a good surface finish and a reasonably fine structure.

As a result of the high viscosity of the aluminium-cobalt alloys in the liquid state, difficulties were experienced in producing sound bars, and in some cases deep "laps" were present on the surface. The aluminium-beryllium alloys required vigorous stirring and a relatively high pouring temperature to overcome the tendency for the beryllium to separate out. Two series of binary aluminium-silicon alloys were cast, one after modification with from 0.04 to 0.08% of sodium, and the second without any modifying treatment.

In general, the alloys were chemically analysed for the contents of the alloying elements only, and no estimation was made of the impurities, which were of a low order and consisted chiefly of iron.

2. SA.24 Alloy in Various Conditions.

Test-pieces were made in a complex alloy, designated SA.24, in various conditions of fabrication. The chemical composition of the alloy was: silicon 11.7, copper 2.14, nickel 1.94, manganese 0.94, magnesium 0.57, cobalt 0.19, titanium 0.05, iron 0.28%, and aluminium the remainder.

(a) *Sand-Cast Bars.*

Standard D.T.D. sand-cast test bars were made in SA.24 alloy to give a comparison between sand-cast and chill-cast materials. The melt was degassed and fluxed with 0.4% of sodium boron fluoride and poured at between 670° and 700° C. into dry sand moulds. The bars were generally sound and had a smooth surface in the cast condition.

(b) *Forged Bars.*

To ensure a fine and uniform cast structure suitable for forging, water-chilled billets were cast in thin-walled moulds of mild steel, approximately 5 in. dia. and 12 in. long. The mould was provided with a sand cup feeder head about 5 in. deep to obtain adequate feeding, and the rate of flow of the water was carefully controlled at about 1.5 gal./min.

Each ingot was machined to give two billets about 4.5 in. dia. and

5 in. long. From the appearance of the machined surface, the material seemed to be relatively free from unsoundness. The billets were homogenized at 480° C. for 72 hr. and were forged at 490° C. down to 1 in. square bar in five stages, with a further treatment of about 1 hr. at the soaking temperature between each stage. The first four forging operations were carried out on a hydraulic press of 60-tons capacity and the final operation on a 10-cwt. pneumatic hammer. After upsetting and squaring, the billet was drawn out at right angles to the original axis and forging continued until bar of 1 in. square cross-section was obtained.

(c) *Sheet Material.*

Two machined billets, each 4.5 in. dia. by 5 in. long, were prepared as described above and forged at right angles to the original axis to give two plates approximately $0.75 \times 3 \times 10$ in. and $1 \times 3 \times 18$ in., respectively. The plates were then machined to give a cross-section of about $\frac{11}{16} \times 2.75$ in., heated to 490° C., and rolled down by stages. In general, the alloy rolled satisfactorily, although some edge-cracking occurred if the percentage reduction per pass was too drastic. This difficulty was mainly overcome by increasing the soaking temperature to about 510° C. for the last two heats and diminishing the percentage reduction employed.

3. *Forged Bars in Aluminium-Nickel and Aluminium-Silicon Alloys and in Complex Alloys Containing Manganese and Nickel.*

In order to determine the effects of manganese and nickel in more complex alloys, a series of melts was prepared containing nominally silicon 12, copper 2, magnesium 0.5, cobalt 0.2, titanium 0.1%, and varying amounts of manganese and nickel. In addition, forged samples were prepared in binary aluminium-nickel and aluminium-silicon alloys to provide a comparison with the results on the corresponding chill-cast alloys.

The complex alloys were degassed with from 0.2 to 0.4% of sodium boron fluoride and cast at from 715° to 760° C.; the rate of water chilling of the ingot was increased with increasing alloying content, and in particular with the nickel content. Degassing of the binary alloys was carried out with 0.4% of Foundry Service No. 7 degasser, and the casting temperatures ranged between 700° and 790° C. The rate of chilling was further increased with the nickel and silicon content of the alloys.

The ingots were $2\frac{3}{4}$ in. dia. by 10 in. long, and were machined to give two small billets each about 2.5 in. dia. and up to 4.0 in. long.

The billets were homogenized at 480° C. for from 24 to 48 hr. and forged from 490° C. to 1 in. square bar in four stages, with a further treatment of about 1 hr. at the soaking temperature between each stage. The billets were upset and squared and drawn out at right angles to the original axis, the first three operations being carried out on a hydraulic press and the final operation on a pneumatic hammer. In general, the materials forged satisfactorily, although the alloys with high contents of manganese and nickel showed a tendency to develop cracks.

4. Heat-Treatment.

The cast and wrought materials in SA.24 alloy were solution heat-treated at 500° C., quenched in cold water, and aged for 16 hr. at 170° C., followed by air cooling. A few specimens of the sheet material were subsequently annealed at 250° C. for 48 hr. prior to testing. The heat-treatments given to the chill-cast bars in the binary and ternary alloys are recorded in the appropriate Table.

5. Method of Testing.

Some initial tensile tests were made using the R.A.E. type of optical extensometer, but owing to the inconsistency of the values obtained for Young's modulus on a given test-piece, all the subsequent work was carried out using a Lamb extensometer of standard design and calibrated at the National Physical Laboratory.

Owing to the relatively low limit of proportionality of some of the alloys in the chill-cast condition, in order to obtain representative values for E , it was necessary to prestrain the material initially to some extension not greater than 0.1% of the gauge length. By strain-hardening in this manner, it was possible to obtain twice the number of readings for each stress-strain curve, and the accuracy of the slope of the line was considerably increased. Four tests were made on each test-piece, the extensometer being removed and replaced after the second run so as to provide values with two settings of the extensometer. The values for Young's modulus obtained by this means were very consistent, and it was considered that the limits of accuracy for the average value of E from four readings was of the order of $\pm 0.1 \times 10^6$ lb./in.²

III.—RESULTS.

The effect, on the Young's modulus of aluminium, of unsoundness, casting stresses, &c., involved in casting from temperatures varying from 720° to 800° C. was not known, and accordingly, chill-cast bars were made in commercially pure aluminium (99.5% purity), one

set being cast from 800° C. and the second from 720° C. In addition, measurements were made on forged and heat-treated bars in the same material.

The average values of Young's modulus for the material cast at 800° C. and 720° C. were 9.6 and 9.3×10^{-6} lb./in.², respectively. The forged and heat-treated bars gave a value of 9.4×10^{-6} lb./in.² It is

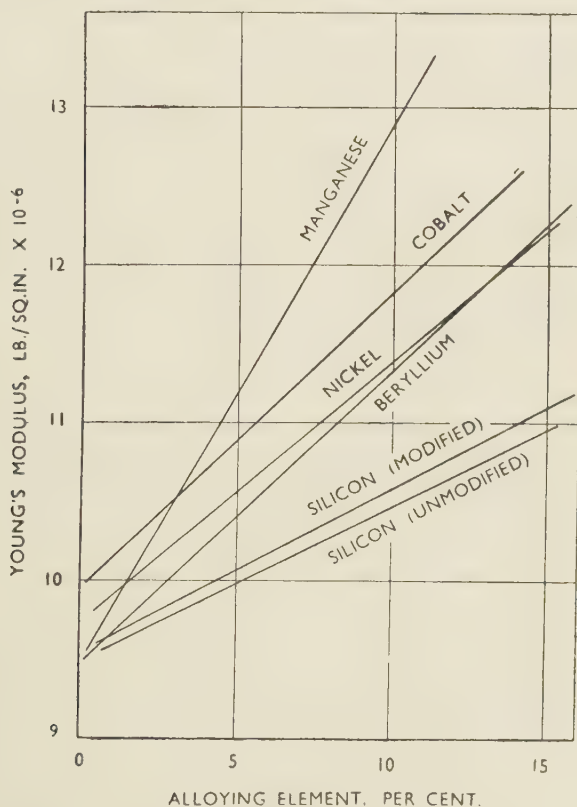


FIG. 1.—Young's Modulus of Binary Aluminium Alloys.

considered that, in view of the small number of tests carried out and the relatively low limit of proportionality for aluminium, there is no significant difference between these results.

The values of E were determined for chill-cast bars in binary aluminium-silicon alloys in both the modified and unmodified conditions. The values obtained are given in Table I, and their relationship with composition is shown in Fig. 1. The modified alloys

TABLE I.—*Values of E for Modified and Unmodified Binary Aluminium-Silicon Alloys in the Chill-Cast Condition.*

Alloy Number.	Silicon, %.	Young's Modulus, lb./in. ² $\times 10^{-6}$.			
		E_1 .	E_2 .	E_3 .	Average E .
P58	5.3	10.15	10.1	10.15	10.15
P58A	5.3	10.2	10.2	10.2	10.2
P59	10.6	10.5	10.5	10.5	10.5
P59A	10.6	10.65	10.65	10.7	10.65
P60	15.2	11.3	11.4	11.3	11.3
P60A	15.2	11.4	11.3	11.2	11.3
P61	19.0	11.6	11.5	11.4	11.5
P61A	19.0	11.9	11.8	11.7	11.8
P62	26.4	12.0	12.0	12.0	12.0
P62A	26.4	12.4	12.4	12.3	12.4
P63	31.0	12.7	12.5	12.5	12.6
P63A	31.0	12.6	12.6	12.9	12.7

Alloys P58A, P59A, P60A : modified with 0.04% sodium.

Alloys P61A, P62A, P63A : modified with 0.08% sodium.

Alloys P58, P59, P60, P61, P62, P63 : unmodified.

gave a slightly higher value of E than the corresponding unmodified alloys, and it is suggested that this was due to an increase in soundness caused by modification. Assuming a linear relationship between E and the composition, the slope of the two curves is the same for the two conditions and represents an increase of 0.104×10^6 lb./in.² in Young's modulus for an increase of 1 wt.-% of silicon. This figure has been used for applying a correction to the experimental value of E in ternary alloys in which the silicon content deviates from the nominal percentage of 12.

Tables II-V give the results of tests on bars in binary aluminium alloys containing varying amounts of beryllium, cobalt, manganese, and nickel, respectively, together with ternary alloys containing approximately 12% silicon and one of the above-mentioned four elements. The results are given for all the alloys both in the as-cast condition and after solution treatment and ageing. The value given in the fifth column in each table is the corrected value of E for a ternary alloy containing 12% silicon after making an allowance for any variation in the silicon content from 12%.

It will be observed that the values of E obtained after heat-treatment agree closely with those given by the material in the as-cast condition. For the as-cast binary alloys, the relationship between the experimental values of Young's modulus and the composition is represented in Fig. 1; the relationship between the "corrected" value of E and composition

TABLE II.—*Values of Young's Modulus for Binary Aluminium-Beryllium and Ternary Aluminium-Beryllium-Silicon Alloys in the Chill-Cast Condition, and After Heat-Treatment.*

Alloy Number.	Chemical Composition, %.		Young's Modulus, lb./in. ² $\times 10^{-6}$.		
			As Cast.		Cast and Heat-Treated.
	Be.	Si.	Average <i>E</i> .	Corrected <i>E</i> .	Average <i>E</i> .
P41	1.21	...	9.7	...	9.7
P42	2.37	...	9.9	...	9.9
P43	2.97	...	10.0	...	10.0
P44	4.25	...	10.3	...	10.4
P45	5.60	...	10.5	...	10.4
P46	1.0	11.6	10.9	11.0	11.0
P47	1.95	11.5	11.3	11.3	11.4
P48	2.97	10.5	11.5	11.6	11.8
P49	3.32	10.6	11.7	11.9	11.5

Heat-treatment : 3 hr. at 500° C.; cold-water quenched; 16 hr. at 175° C.

TABLE III.—*Values of Young's Modulus for Binary Aluminium-Cobalt and Ternary Aluminium-Cobalt-Silicon Alloys in the Chill-Cast Condition, and After Heat-Treatment.*

Alloy Number.	Chemical Composition, %.		Young's Modulus, lb./in. ² $\times 10^{-6}$.		
			As Cast.		Cast and Heat-Treated.
	Co.	Si.	Average <i>E</i> .	Corrected <i>E</i> .	Average <i>E</i> .
P30	1.20	...	10.2	...	9.6
P31	2.08	...	10.3	...	9.8
P32	3.10	...	10.5	...	10.1
P33	4.01	...	10.6	...	10.4
P34	6.57	...	11.3	...	11.4
P36	1.17	11.55	11.4	11.4	11.5
P37	2.15	10.00	11.6	11.8	11.9
P38	3.13	12.10	11.9	11.9	12.0
P39	5.10	11.40	12.2	12.2	12.3
P40	8.00	13.15	13.2	13.1	13.1

Heat-treatment : 3 hr. at 500° C.; cold-water quenched; 16 hr. at 175° C.

TABLE IV.—*Values of Young's Modulus for Binary Aluminium-Manganese and Ternary Aluminium-Manganese-Silicon Alloys in the Chill-Cast Condition, and After Heat-Treatment.*

Alloy Number.	Chemical Composition, %.		Young's Modulus, lb./in. ² $\times 10^{-6}$.		
			As Cast.		Cast and Heat-Treated.
	Mn.	Si.	Average <i>E</i> .	Corrected <i>E</i> .	Average <i>E</i> .
P19	1.67	...	10.2	...	10.0
P20	3.45	...	10.6	...	10.5
P21	5.16	...	11.2	...	11.1
P22	8.40	...	12.3	...	12.3
P23	10.00	...	13.1	...	13.0
P24	14.56	...	14.2	...	14.2
P25	1.58	13.7	11.6	11.4	11.7
P26	3.03	14.0	12.0	11.8	12.2
P27	4.86	12.95	12.6	12.5	12.7
P28	7.70	13.85	13.5	13.3	13.5
P29	9.80	14.5	14.0	13.8	14.0

Heat-treatment : 3 hr. at 500° C.; cold-water quenched; 16 hr. at 175° C.

for the ternary alloys is shown in Fig. 2. From the slope of these curves, the values in Table V have been obtained for the effect on the Young's modulus of 1 wt.-% addition of each of the alloying elements investigated.

TABLE V.

Element.	Increment in <i>E</i> for 1 wt.-% of Element, lb./in. ² $\times 10^{-6}$.	
	In Binary Alloys.	In Ternary Alloys with 12% Silicon.
Beryllium . . .	0.188	0.340
Cobalt	0.186	0.250
Manganese . . .	0.340	0.307
Nickel	0.164	0.162
Silicon	0.104	...

From Table V, it will be observed that, with the exception of nickel, the increase in Young's modulus for 1 wt.-% of alloying element is different in binary alloys to that obtained in the ternary alloys containing silicon. It is possible that, for elements which form ternary constituents with aluminium and silicon in the aluminium-rich alloys, the introduction of a new phase would result in different rates of increase of *E* for binary and ternary alloys. This would explain the

agreement between the two increments for nickel, which does not form a ternary constituent, and the difference between the values for manganese, which forms two ternary aluminium-manganese-silicon phases.

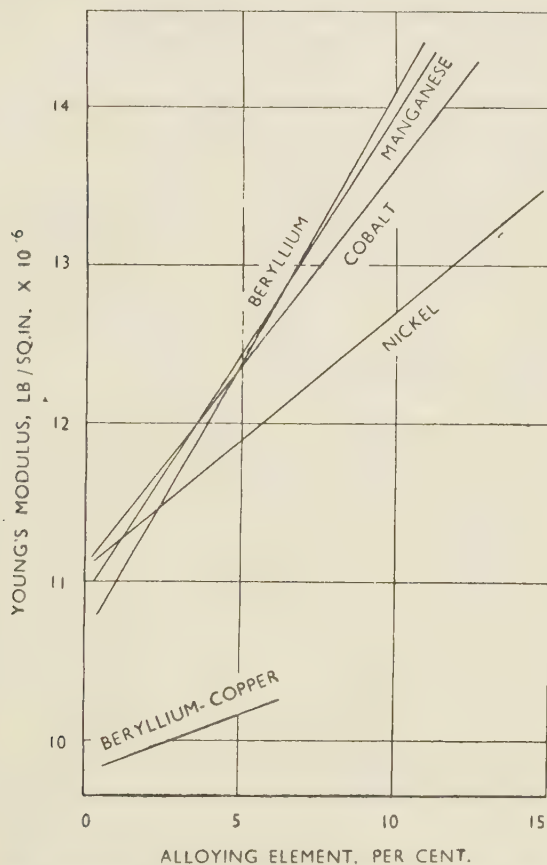


FIG. 2.—Young's Modulus of Ternary Aluminium Alloys Containing 12% Silicon, and of Ternary Aluminium-Beryllium-Copper Alloys Containing 2.4% Copper.

According to published information, however, no ternary phase is formed in either the aluminium-cobalt-silicon or the aluminium-beryllium-silicon system, so that the difference obtained for these elements is not easily explained. In the aluminium-beryllium-silicon system, a binary phase between beryllium and silicon, similar to Mg_2Si , is possible, but, as far as is known, has never been identified. Accordingly, the binary and ternary alloys with beryllium and cobalt were

microscopically examined at the Royal Aircraft Establishment in the hope of finding some feature which would afford a possible explanation, but none was obtained (see the Section on examination of micro-structure).

It appears, therefore, that the value of Young's modulus obtained for complex alloys may also depend upon some factor other than the elastic constants of the constituents present in the structure. No generalized statement can be made until the results are available on alloys containing chromium, copper, iron, magnesium, molybdenum, titanium, tungsten, vanadium, &c.

TABLE VI.—*Values of Young's Modulus for Binary Aluminium-Nickel and Ternary Aluminium-Nickel-Silicon Alloys in the Chill-Cast Condition, and After Heat-Treatment.*

Alloy Number.	Chemical Composition, %.		Young's Modulus, lb./in. ² $\times 10^{-6}$.		
			As Cast.		Cast and Heat-Treated.
	Ni.	Si.	Average <i>E</i> .	Corrected <i>E</i> .	Average <i>E</i> .
P 7	0.99	...	10.0	...	9.8
P 8	2.89	...	10.2	...	10.4
P 9	5.00	...	10.5	...	10.5
P10	7.56	...	11.1	...	11.0
P11	10.00	...	11.3	...	11.2
P12	14.65	...	12.1	...	12.4
P13	1.03	12.8	11.4	11.3	11.2
P14	3.00	12.33	11.6	11.5	11.5
P15	4.90	10.4	11.6	11.8	11.8
P16	8.85	14.8	12.8	12.6	13.0
P17	10.1	12.6	13.1	13.0	13.2
P18	15.8	13.4	13.4	13.3	13.3

Heat-treatment : 3 hr. at 500° C.; cold-water quenched; 16 hr. at 175° C.

TABLE VII.—*Average Values of Young's Modulus for Forged and Heat-Treated Aluminium-Nickel Alloys.*

Alloy Number.	Nickel, %.	Average <i>E</i> , lb./in. ² $\times 10^{-6}$.
MY17	2.68	9.9
MY18	5.0 *	10.3
MY19	7.35	11.0
MY20	10.0 *	11.4

Heat-treatment : 5 hr. at 520° C.; cold-water quenched; 16 hr. at 180° C.

* Nominal chemical composition.

The results obtained for binary aluminium-nickel and aluminium-silicon alloys in the forged condition are shown in Tables VII and VIII, respectively. With the exception of the alloy containing 21% silicon which showed some cracking, the values of Young's modulus obtained on the forged bars agree within experimental error with the results quoted for the corresponding alloys in the chill-cast condition.

TABLE VIII.—Average Values of Young's Modulus for Binary Aluminium-Silicon Alloys in the Forged Condition.

Alloy Number.	Silicon, %.	Average E , lb./in. ² $\times 10^{-4}$.
MY13	5.0 *	10.1
MY14	10.25	10.8
MY15	15.0 *	11.2
MY16	21.3	11.4

Bar MY16 showed cracks.

* Nominal chemical composition.

The values of Young's modulus for the ternary aluminium-beryllium-copper and quaternary aluminium-beryllium-copper-silicon alloys are given in Table IX, and the relationship between E and composition

TABLE IX.—Values of Young's Modulus for Ternary Aluminium-Beryllium-Copper and Quaternary Aluminium-Beryllium-Copper-Silicon Alloys in the Chill-Cast Condition, and After Heat-Treatment.

Alloy Number.	Chemical Composition, %.			Young's Modulus, lb./in. ² $\times 10^{-4}$.		
				As Cast.		Cast and Heat-Treated.
	Be.	Cu.	Si.	Average E .	Corrected E .	Average E .
P51	1.10	2.40	...	9.9	...	9.8
P52	2.18	2.52	...	9.9	...	10.0
P53	3.15	2.46	...	10.0	...	10.1
P54	4.17	2.35	...	10.1	...	10.1
P55 *	1.07	2.48	10.5	10.8	10.9	10.9
P56	2.27	2.47	10.7	11.5	11.6	11.6
P57 *	3.27	2.35	10.8	11.2	11.3	11.3

Heat-treatment: 3 hr. at 500° C.; cold-water quenched; 16 hr. at 175° C.

* Slightly porous bars.

is shown in Fig. 2. The bars, P55 and P57, in the quaternary alloys were slightly porous, so that the values of E obtained were somewhat lower than would have been expected for sound material. From the

results for the ternary alloys, an increment in E of 0.08×10^6 lb./in.² is obtained for 1 wt.-% increase in beryllium. This value is considerably lower than those obtained for the effect of 1% beryllium in binary aluminium-beryllium and ternary aluminium-beryllium-silicon alloys. The most probable explanation of this is that the elastic properties of the ternary aluminium-beryllium-copper constituent occurring in these alloys (see Section on examination of microstructure) are lower than those of beryllium itself.

TABLE X.—*Values of E for SA.24 Alloy in Various Conditions.*

Condition.	Test-Piece Number.	Young's Modulus, lb./in. ² $\times 10^{-6}$.		
		E_1 .	E_2 .	Average.
Sand-cast bar, f.h.t.	1	11.5	11.5	11.5
	2	11.5	11.5	
	3	11.5	11.5	
Chill-cast bar, f.h.t.	1	11.8	11.8	11.73
	2	11.7	11.7	
	3	11.7	11.7	
Forged bar, 1 in. square, f.h.t.	1	11.9	12.0	11.95
	2	11.9	11.9	
	3	11.9	12.0	
	4	12.0	12.0	
f.h.t. and annealed 48 hr. at 250° C.	5	12.1	12.1	12.0
	6	12.0	11.9	
3 S.W.G. sheet, f.h.t.	1	11.9	12.0	11.9
	2	11.9	11.9	
	3	11.9	11.9	
20 S.W.G. sheet, f.h.t.	1	11.4	11.5	11.5 *
	2	11.5	11.5	
	3	11.1	11.1	11.1 *
	4	11.1	11.1	

f.h.t. : heated 5 hr. at 520° C.; cold-water quenched; 16 hr. at 175° C.

Nominal composition of SA.24 : silicon 12.0, copper 2.0, nickel 2.0, manganese 1.0, magnesium 0.5, cobalt 0.2, titanium 0.05%, and aluminium remainder.

* Similar results were obtained on two further test-pieces.

The results obtained on the experimental high-silicon alloy SA.24 in various conditions of fabrication are given in Table X. It will be observed that, with the exception of the sand-cast material and the 20-S.W.G. sheet, the values of E obtained on the alloy in the various conditions are the same, within experimental error. The somewhat lower value obtained for the sand-cast material is probably due to

porosity, which is usually present in sand castings, whereas the reasons for the abnormally low values for the 20-S.W.G. sheet are considered to be related to the difficulties of obtaining accurate extensometer measurements on thin, flat test-pieces. Further, for the 20-S.W.G. material, four test-pieces gave a consistent value of E of about 11.5×10^6 lb./in.², whereas four further test-pieces gave a consistent value of about 11.1×10^6 lb./in.². This difference might be explained by the fact that the rate of reduction during rolling was greater for some samples than others, and that the samples with the higher value of E received less cold work than those with the lower E . Similar effects have been observed by Japanese workers⁴ for aluminium, copper, and nickel.

TABLE XI.—*Tensile Properties of Complex Aluminium Alloys Containing Varying Amounts of Manganese and Nickel.*

Alloy Number.	* Chemical Composition, %.		0.1% Proof Stress, tons/in. ²	Maximum Stress, tons/in. ²	Elongation, % on 4√A.	Young's Modulus, lb./in. ² × 10 ⁻⁶ .	
	Mn.	Ni.				Observed.	Calculated.
MY 1	...	1.97	21.8	28.2	8	11.6	11.1(5)
MY 2	...	3.55	21.7	27.0	3	12.0	11.4
MY 3	...	5.10	23.2	27.8	2	12.3	11.6(5)
MY 4	0.62	...	21.5	27.9	4	11.2	11.0
MY 5	1.81	...	20.5	24.2	3	11.9	11.4
MY 6	0.82	2.10	19.0	25.6	4	11.8	11.4
MY 7	0.88	3.60	20.4	26.8	3.5	12.2	11.6(5)
MY 8	0.82	5.10	20.7	26.1	2	12.5	11.9
MY 9	1.64	2.00	18.3	23.2	3.5	12.1	11.6(5)
MY10	0.94	3.66	19.9	23.3	2	12.7	11.7
MY11	0.57	2.06	18.2	21.6	4	11.95	11.3

* N.B.—With the exception of alloy MY11, which was copper-free, all the alloys, in addition to the above elements, contained nominally silicon 12, copper 2, magnesium 0.5, cobalt 0.2, iron about 0.3, and titanium 0.1%.

Table XI gives the values of 0.1% proof stress, maximum stress, percentage elongation, and Young's modulus on the complex series of alloys containing various amounts of manganese and nickel. Also included in the Table are the values of E for each alloy, calculated from the composition, using the increments in E for cobalt, manganese, and nickel found for the ternary alloys. It has been assumed that the effect of iron is of the same order as that of cobalt, and that the copper and magnesium present have no appreciable effect on the modulus as indicated by statistical analysis on the results of the original series of alloys. It will be observed that the experimental value of E is in each case greater than that obtained by calculation, and, although no

explanation of this can be given, it is noteworthy that the effects of manganese and nickel are at least as great in the complex alloys as in the binaries. The results indicate that in order to obtain high values of proof stress and maximum stress combined with high E , it is necessary to make a compromise between the contents of nickel, manganese, and copper; with further work, it should be possible to develop alloys with useful properties for many applications.

IV.—DEVELOPMENT OF ALLOYS FOR COMMERCIAL APPLICATION.

No attempt has been made to develop alloys at the Royal Aircraft Establishment having a high Young's modulus combined with high mechanical properties, but an investigation⁵ by W. Stalker is in progress to determine the general properties of a selected complex alloy previously developed² by N. Dudzinski and E. I. Brimelow. The alloy was designated SA.23 and was of the following composition: silicon 12.0, copper 1.75, nickel 1.0, manganese 2.0, magnesium 0.47, cobalt 0.28, titanium 0.09, iron 0.25%, and aluminium the remainder.

Water-chilled billets, 5 in. dia. were cast, forged into 1 in. square bar, and rolled to 16-S.W.G. strip. The material was solution heat-treated for 2 hr. at 520° C., quenched in cold water, and aged for 17 hr. at 170° C. The specific gravity of the worked material was 2.8, which is only slightly higher than that of commercial aluminium alloys for structural use. Tensile tests gave results in the ranges shown in Table XII.

TABLE XII.

Condition.	0.1% Proof Stress, tons/in. ²	Max. Stress, tons/in. ²	Elongation, %.	$E \times 10^{-6}$, lb./in. ²
Forged and heat-treated .	19.8–20.5	23.5–25.2	1.5–2.5	12.2–12.4
Rolled and heat-treated .	19.8–20.1	24.4	2.0–3.0	10.8–11.6

The values of proof stress, maximum stress, and elongation are all markedly lower than those obtained on extruded spars and sheet materials in the standard aluminium alloys at present in use. The values of E for the forged test-pieces were all in the range $12.2\text{--}12.4 \times 10^6$ lb./in.², but those for the sheet samples showed considerable scatter, which has not been satisfactorily explained. These results may be regarded as confirming the tendency, referred to above in connection with SA.24 alloy, for cold reduction to reduce the modulus. The determination of E on thin test-pieces is a difficult operation, and it is probable that the use of stabilizing plates, similar to those used by Grover, Munro, and Chalmers, would lead to more consistent results.

Further work is in progress to investigate methods of determining the elastic modulus of sheet materials.

Wohler fatigue tests on the forged material gave fatigue limits of 9.9, 9.6, and 9.3 tons/in.², respectively, for endurances of 10, 20, and 50 million cycles. These results are of the same order as those obtained on aluminium alloys to British Standard Specifications L.40 and L.39. Corrosion and stress-corrosion tests are not yet complete, but the results to date indicate that the alloy has satisfactory resistance to these forms of attack.

V.—EXAMINATION OF MICROSTRUCTURE.

Sections from various alloys in the chill-cast condition were microscopically examined to determine the phases which occurred and their mode of occurrence. Sections were polished both by mechanical means and electrolytically using the solution recommended by De Sy and Haemers.

The structure of the binary aluminium–beryllium alloys consisted of primary beryllium particles in the eutectic matrix of α solid solution and beryllium. The eutectic structure was generally fine and similar in appearance to that of aluminium–silicon alloys. With increase in the beryllium content, both the primary particles and the eutectic formations became more coarse. The ternary aluminium–beryllium–copper alloys contained, in addition to large massive particles of beryllium, two copper-bearing phases, considered to be a ternary aluminium–beryllium–copper constituent and CuAl_2 , respectively. The ternary constituent occurred in association with the massive beryllium particles and also as small, discrete separations in the grains. The CuAl_2 phase was present only as small separations in the grain boundaries. No ternary aluminium–beryllium–silicon phase was detected in the aluminium–beryllium–silicon alloys, and the microstructure consisted of somewhat coarse aluminium–silicon eutectic, very fine aluminium–beryllium eutectic, and large massive particles of primary beryllium.

The Co_2Al_9 constituent in the binary aluminium–cobalt alloys varied from small plates and needles in the 1% alloy to coarse, chain-like formations in the 8% alloy. The structure of the ternary aluminium–cobalt–silicon alloys consisted essentially of primary particles of silicon and cobalt aluminide with some coarse separations of aluminium–silicon eutectic. No particles were detected which would correspond to a ternary aluminium–cobalt–silicon constituent.

The binary aluminium–manganese alloys of relatively high alloying content contained some MnAl_4 particles, but for the lower contents of

manganese the structure consisted essentially of solid solution and MnAl_6 . The three constituents identified for the ternary aluminium-manganese-silicon alloys were aluminium, primary silicon, and ternary aluminium-manganese-silicon.

VI.—DISCUSSION.

Before discussing the interpretation of the experimental results, it is appropriate to consider the accuracy of the data. The conclusions to be drawn will be based on the linearity of the composition-modulus relationships and on the slopes of the lines obtained. Experimental error may have arisen both in the composition and in the modulus determination, and it is difficult to examine in detail the possible causes and amount of these errors. The fact that the data are self-consistent, and that the points all fall close to straight lines, is strong evidence that the errors are small and that the results are not vitiated by experimental error.

As no effort was directed towards the development of alloys combining high strength with high Young's modulus, little attention has been paid to the tensile properties of the alloys investigated. Such a study should form the next stage of the work and be made under conditions in which the casting and hot-working properties of the alloys can be fully assessed in relation to their mechanical properties. It is apparent that, while an increase in the amounts of certain alloying elements such as nickel and manganese results in an increase in Young's modulus, the hot-working properties of the alloys correspondingly deteriorate. Further, although manganese increases the value of E and the introduction of copper enables heat-treatment to be carried out for the development of high proof stresses and a high tensile strength, it is necessary to make a compromise between the contents of copper and of manganese to facilitate fabrication of the alloys and to ensure that both the tensile and elastic properties obtained are of a useful order. However, from the results on the complex alloys, quoted in Table XI, and on other alloys investigated at the Royal Aircraft Establishment, it is reasonable to expect that alloys having the following minimum tensile properties, combined with satisfactory hot-working properties, can be developed.

0.1% proof stress, tons/in. ²	21.0
Maximum tensile stress, tons/in. ²	28.0
Elongation on $4\sqrt{A}$, %	4.0
Young's modulus, 10^6 lb./in. ²	12.0

It is considered that, as a result of further work, it will be possible to make small additions of certain elements to give alloys with a Young's

modulus greater than 12.0 without any marked lowering of the tensile properties and ductility quoted above.

It may be concluded from the results that the elements manganese, beryllium, cobalt, nickel, and silicon are all effective in increasing the Young's modulus by amounts that are proportional to their concentrations; the effects of these alloying elements are to some extent interdependent, and the total increase in modulus of a complex alloy is generally not less than the sum of the effects to be expected of each separately. The general conclusion is that the intermetallic-compound phases of aluminium with manganese, cobalt, and nickel have relatively high moduli and that the modulus of the composite material is in each case due to the separate contributions of the solid-solution matrix and the other constituents that are present. It has been shown by Middleton ⁶ that the modulus of such an aggregate can be calculated approximately from the moduli of the respective constituents, and that the relationship to be expected is approximately linear for relatively low concentrations such as have been used in the experiments under discussion.

Although no theoretical work appears to be available on the factors controlling the elastic moduli of intermetallic compounds, the authors consider that the heat of formation of a compound should be positively correlated with the elastic modulus. This view has also been put forward in a private communication by Dr. Raynor, who points out that since the increase in modulus will be determined by the modulus of the compound and by the relative volume of the compound present, the aluminium content of the compound phase should be as high as possible in order that a given weight percentage of alloying element should have the maximum effect. Figures quoted by Dr. Raynor show that these views are supported by the experimental evidence available.

In conclusion, it should be emphasized that no claim is made to have developed alloys of practical value with enhanced Young's modulus values; it has, however, been shown how a significant, and perhaps useful, increase in this property can be obtained.

VII.—FURTHER DEVELOPMENTS.

It is believed that other elements, such as chromium, iron, molybdenum, titanium, tungsten, and vanadium, which form intermetallic constituents with aluminium, also have beneficial effects on the Young's modulus of aluminium and its alloys. The properties of binary and more complex alloys containing these elements should be investigated

to determine the magnitude of the effect of each and to obtain a better understanding of the interaction of various alloying elements. Before a complete theoretical understanding of the Young's modulus of complex alloys can be obtained, it will be necessary to devise a more accurate method for the determination of E , combined with a very close control of composition throughout the samples used. By this means, it should be possible to investigate the elastic properties of solid solutions and the effect of heat-treatment, &c.

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APPENDIX.

THE MODULI OF ALUMINIUM ALLOYS IN TENSION AND COMPRESSION.

By S. F. GROVER, W. MUNRO, and B. CHALMERS.

I.—INTRODUCTION.

In connection with the significance of Young's modulus in aircraft design, it was desired to investigate the compressive properties of sheet materials. These properties were at first investigated by the single-thickness method described by Paul, Howell, and Grieshaber,⁷ but it was found that this method gave values for the compression modulus approximately 6% higher than the modulus obtained in tensile tests. As this was contrary to the fundamental theory of elasticity, other methods than rollers for stabilizing the sheet were tried, but these gave the same discrepancy. In view of this, it was decided to make tests in which the compression modulus and tension modulus could be determined without the extensometer being removed from the test-piece or the test-piece being removed from the testing machine.

II.—TEST PROCEDURE.

1. *Bar Material.*

With the bar material to British Standard Specification 6L1, a test-piece of the shape shown in Fig. 3 was used. This test-piece was loaded

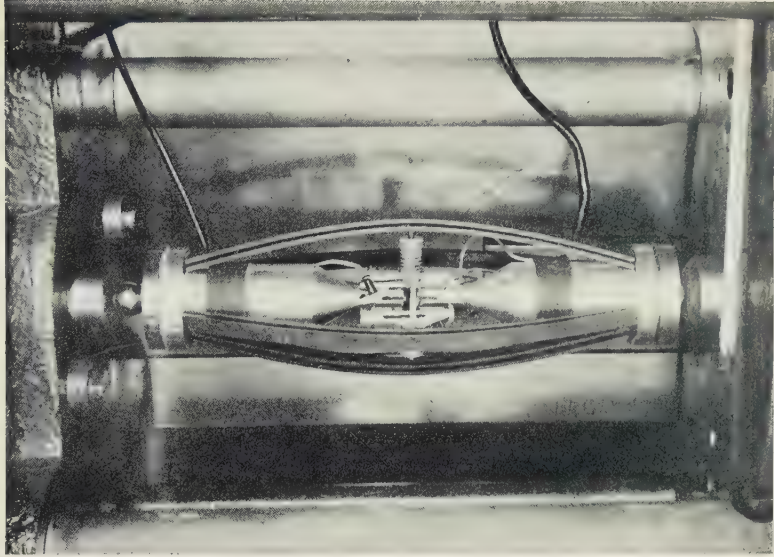


FIG. 5.—Bar Test-Piece with Strain Gauges and Lamb Extensometers.

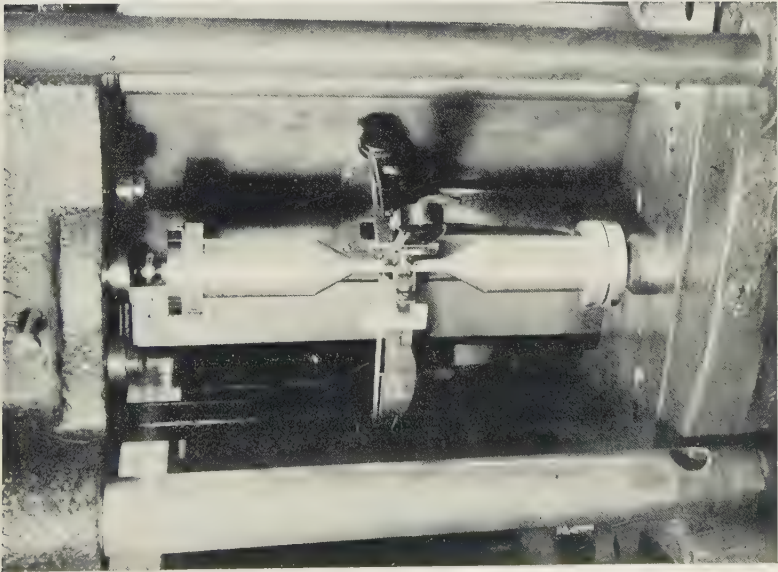


FIG. 4.—Bar Test-Piece with Huggenberger Extensometers.



FIG. 8.—Sheet Test-Piece under Tensile Load.

axially in compression up to a stress of 5 tons/in.² on the reduced section, the axially of load being measured by three Huggenberger strain-gauges, placed at intervals of 120° around the bar, as shown in Fig. 4 (Plate XXII). At this stage, three wire resistance strain-gauges were stuck on to the test-piece, also at 120° intervals, and the load then removed. The wire resistance gauges were stuck on with the test-piece under compression so that they would never be subjected to compression themselves.

A tensile load was applied to the test-piece by means of eight buckled struts, as shown in Fig. 5 (Plate XXII), the axially of loading being checked by the wire resistance gauges. The load obtained by the buckled struts was approximately 1100 lb., equivalent to 3.9 tons/in.² on the reduced section. The test-piece, with buckled struts assembled, was placed in a compression testing machine and an axial compressive load applied, the strains being measured by means of the wire resistance gauges and also a Lamb roller-type extensometer. The compressive load applied by the testing machine was increased in steps up to 0.9 ton, strain readings being taken for each step. Under these conditions, the stress on the reduced section of the test-piece decreased from 3.9 tons/in.² tension, with no compressive load applied by the machine, through zero, to 3.3 tons/in.² compression, when the testing machine was applying 0.9 ton compression. Typical results of this test are shown in Fig. 6.

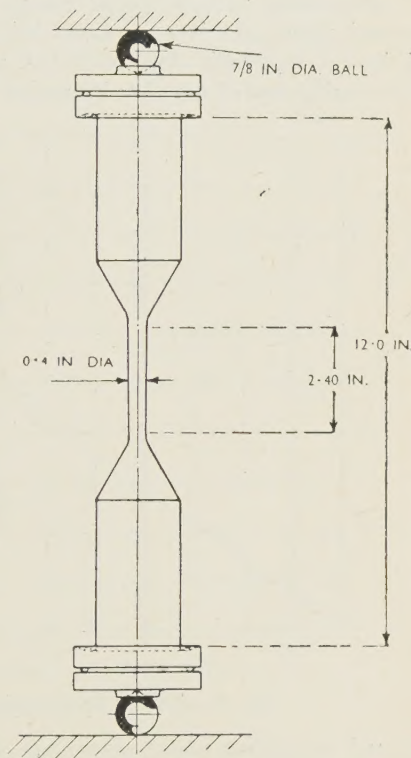


FIG. 3.—Tension-Compression Test-Piece.

2. Sheet Material.

The test-piece used for the sheet material is shown in Fig. 7. End pieces were clamped to the test-piece as illustrated in Fig. 8 (Plate

XXIII) in order to accommodate the six buckled struts for applying the tensile load. The enlarged end of the test-piece just protruded beyond the end plates, so that the compressive load was applied directly to the sheet material, so obviating the necessity of reversing the direction of loading through the hole in the end of the test-piece.

The centre portion of the test-piece was supported between two ground plates, the inner surfaces being coated with a single layer of $\frac{1}{16}$ -in.-dia. steel balls close packed in petroleum jelly. The two plates were spring loaded together by means of a series of leaf springs. The

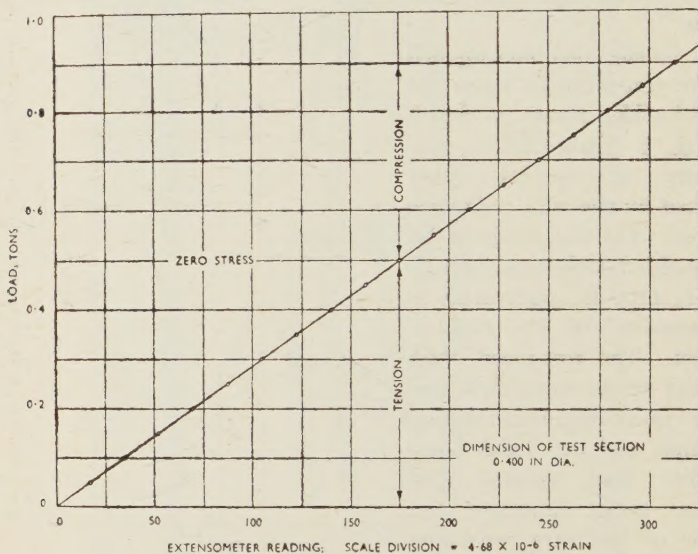


FIG. 6.—Tension-Compression Test-Bar Material.

ground supporting plates were made to intrude into recesses cut into the end pieces in order to give rigidity at the joint, and the reduced section of the test-piece protruded approximately $\frac{1}{32}$ -in. on both sides of the stabilizing plates to allow the Lamb roller-type extensometer to be attached.

The test procedure was similar to that used for bar material, except that no attention was paid to the measurement of axiality of loading.

III.—TEST RESULTS.

The result obtained from the test on the bar material is shown in Fig. 6, and on visual examination no difference between the tensile and compressive Young's modulus can be detected. Similar results were

obtained from the test on sheet material, and no difference in modulus between tension and compression could be detected by visual examination of the curve.

Statistical-curve fitting with the test results indicated that there

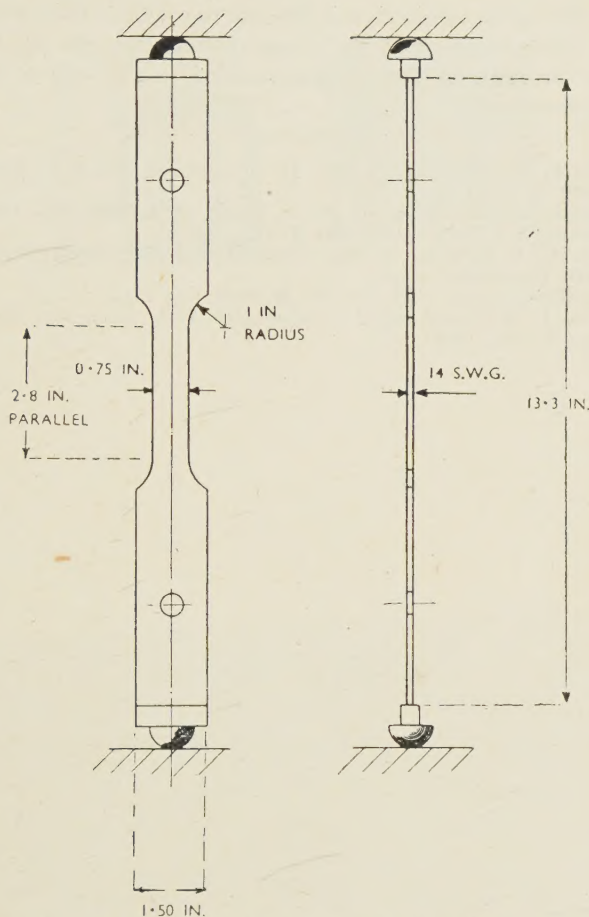


FIG. 7.—Tensile-Compression Tests on Sheet.

was some significant deviation from the straight line, and that the deviation was of the right form to be accounted for by the increase in gauge length and decrease in cross-section. However, this effect was of an order of less than 1% , and it was thought that greater care would have to be taken in the extensometry and load application before such

results could be accepted as significant. In consequence, no details of the curve fitting are given in this report.

IV.—CONCLUSION.

From the results obtained, it is felt safe to conclude that, within the limits of accuracy required for stress calculations, the modulus in tension is the same as that for compression in the case of the two materials concerned.

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